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HEAT-TREATMENT OF METALS

PEACE PUBLISHERS
MOSCOW

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CONVENTIONAL SYMBOLS

Value	Letter symbol	Unit of measurement
Tensile strength (ultimate strength)	σ_t	kg/mm²
Yield point (yield strength), physical or conventional, defined as the stress that will cause the permanent set in a specimen, equal to 0.2 per cent of its gauge length. The yield strength is characteristic of the metal elastic properties	σ_y	kg/mm²
Per cent elongation, when the ratio of the specimen gauge length l to its diameter d is equal to $l:d=5$	σ_5	%
Per cent reduction of area (contraction)	ψ	%
Impact strength, determined by testing the specimens of Menager type	a_i	kgm/cm²
Brinell hardness number	H_B	kg/mm²
Rockwell "C" hardness number (C scale)	R_C	—
Temperature	t	°C
Absolute temperature	T	°C

For the above-listed properties and values no units of measurement are indicated in the tables and figures.

PART ONE

TRANSFORMATIONS IN STEEL DURING ITS HEAT-TREATMENT

Chapter I

GENERAL DATA ON STEELS

1. Classification of Steels

Steels are alloys, the essential ingredients of which are: a) iron, and b) carbon in quantity not exceeding 2 per cent.

High-alloy complex products, composed of iron, chromium, nickel and other elements, in which the maximum content of carbon is limited to very low values (0.10 to 0.12 per cent), are also called steels. In these steels, carbon may be considered essentially as an undesirable impurity, but the number of such steels is very small. For the overwhelming majority of steels carbon is the second essential ingredient after iron.

Steels are also classified in two groups, according to the uses for which their properties fit them, i. e., structural steel, and tool steel. Structural steels with special physical and physico-chemical properties may be subdivided into separate groups embracing corrosion-resistant, acid-resisting, non-scaling, heat-resisting, electrical, magnetic, low-magnetic (non-magnetic), and wear-resisting steels.

Tool steels differ from structural ones by their higher carbon content. The carbon content in structural steels usually does not exceed 0.7 per cent; the majority of tool steels contains more than 0.7 per cent carbon, but in some highly alloyed tool steels the proportion of carbon may be less than this figure.

All steels are classified according to their *chemical composition* as *carbon steels* or *alloy steels*. Carbon steels are those containing only two elements: iron and carbon. Such elements as silicon, manganese, phosphorus and sulphur, contained in each steel, including those of the carbon type, are considered impurities, not ingredients. Steels containing up to 0.2 per cent carbon are listed by the U.S.S.R. State Standard in the group of low-carbon steels; those with a carbon range from 0.25 to 0.6 per cent in the group of medium-carbon steels, and steels with a carbon content above 0.6 per cent in the group of high-carbon steels.

Alloy steels are those which, in addition to the essential ingredients, i. e., iron and carbon, contain special alloying elements — chromium, nickel, tungsten, molybdenum, vanadium, aluminium, titanium, etc.

In some cases silicon and manganese are considered the alloying elements, whereas chromium, nickel and other typical alloying elements may be considered impurities.

The silicon content, as an impurity, in plain carbon and in most low-alloyed steels should not exceed 0.4 per cent. In some high-alloy steels the maximum silicon content is 0.6 per cent. In steels alloyed with silicon the proportion on this element, in general, exceeds 0.8 per cent. As is clear from the above data, it is impossible for all grades of steel to state precisely up to what proportion silicon is to be considered as an ordinary impurity, and above what figure it becomes an alloying element. But conditionally one may accept the following figures as valid for the majority of steels; silicon up to 0.8 per cent is considered an impurity, and over 0.8 per cent — an alloying element.

Similarly, if the whole totality of steel is considered, it cannot be precisely stated up to what proportion manganese is an impurity, and above what figure it becomes an alloying element. One can only accept, in general, that manganese up to 1 per cent is an impurity, and over that figure—an alloying element.

Almost all types of carbon steel contain chromium and nickel. These elements become ingredients of steel due to the usage of scrap alloy steel in charges for melting furnaces. State Standards for plain carbon steels specify the permissible contents of chromium and nickel as not exceeding 0.30 per cent of each. In chromium alloyed steels the minimum content of this element is as much as 0.40 per cent. The minimum proportion of nickel in steels alloyed with this element is 0.50 per cent.

For any grade of steel it can very easily be determined whether a given element is an impurity or an alloying element if a standard steel grade designation is considered. But this is possible only in cases when the steel grade designation is composed of the letter symbols for elements. Impurities may also be distinguished from alloying elements by the fact that in specifications of the chemical composition of steel only the upper limits of content of the former are indicated (e. g., $\leqslant 0.30$ per cent), whereas for the latter the lower limits are indicated as well (e. g., 0.50 to 0.90 per cent). This rule, however, does not apply to silicon and manganese, for which elements, if they are indeed considered as impurities, both the upper and lower limits are indicated in specifications. For instance, the silicon content of most carbon steels is accepted as ranging from 0.17 to 0.37 per cent.

This is caused by intentional adding of silicon and manganese (in the case of rimmed steel only manganese is added) to the steel for its deoxidation; therefore presence of these elements in steel testifies to its having been deoxidised.

By combining these two classifications (namely, according to purpose and to chemical composition) in one common classification, the following five groups of steel have been obtained:

1) carbon structural steels;
2) carbon tool steels;
3) alloyed structural steels;
4) alloyed tool steels;
5) highly alloyed structural steels with special physical and physico-chemical properties.

Apart from the classification of steels according to their purposes and chemical composition, they are also classified according to their quality characteristics as:

1) ordinary quality steels;
2) higher quality steels;
3) quality steels;
4) high-quality steels.

The distribution of the above five groups of steel according to quality is given in Table 1.

Table 1

Classification of Steels According to Quality and Methods of Manufacture

Ordinary and higher quality steels	Quality steels	High-quality steels
Carbon structural steels		
	Carbon tool steels	
	Alloyed structural steels	
		Alloyed tool steels
		Highly alloyed structural steels with special properties
Open-hearth steels Bessemer steels	Open-hearth steels Electric furnace steels	Electric furnace steels Open-hearth steels

According to State Standards alloyed steels are classified as: *low-alloy* steels, if the total (summary) amount of alloying elements contained does not exceed 2.5 per cent; *medium-alloy* steels, if

this amount is in the range of 2.5 to 10 per cent; and *high-alloy* steels, if it exceeds 10 per cent.

The qualitative class of steel is determined by the degree of its purity with regard to harmful impurities and foreign non-metallic matters (inclusions), the uniformity of its chemical composition, and, as a consequence, the higher values of its mechanical and physico-chemical properties. The lower the permissible contents of sulphur and phosphorus in steel, the higher is the class of its quality (see Table 2).

Table 2

Permissible Contents of Sulphur and Phosphorus in Steel

Group of steel	Permissible content in per cent	
	Sulphur	Phosphorus
Carbon structural steels of ordinary quality		
open-hearth steels	0.055	0.050
Bessemer steels	0.060	0.080
Higher quality and quality steels	0.040	0.040
Carbon tool steels		
quality steels	0.030	0.035
high-quality steels	0.020	0.030
Alloy structural steels		
quality steels	0.040	0.040
high-quality steels	0.030	0.035
Alloy tool steels	0.030	0.030
High-alloy structural steels with special properties	0.030	0.035

In analysing figures given in Table 2 a question may often arise: is it really possible that such a negligible difference (0.01 per cent) in sulphur and phosphorus contents of quality and high-quality steels can bring about an appreciable change in properties of these steels? Such a change, of course, is caused not only by the difference in sulphur and phosphorus contents. The sulphur and phosphorus proportions are factors characteristic of purity of steel, factors that can easily be determined by chemical analysis. As a rule, it may be accepted that if the contents of sulphur and phosphorus in steel are low it will also be cleaner with regard to other harmful impurities (oxygen, nitrogen, etc.), which are difficult to determine analytically, and especially with respect to solid non-metallic inclusions.

The specified purity of steel with regard to harmful impurities and non-metallic inclusions can be achieved by proper selection of charging materials and of method of manufacture. The highest purity is obtained by melting steel in electric furnaces, the lowest purity—by making it in Bessemer converters. On the other hand, electric furnace steel is the most expensive, whereas Bessemer steel is the cheapest. Selection of manufacturing methods for steels of various quality classes is made on the basis of these considerations.

Structural steels of *ordinary quality* are made in open-hearth furnaces or in Bessemer converters. Higher quality steels are almost exclusively made in open-hearth furnaces. *Quality* steels are, in general, manufactured in open-hearth furnaces and, sometimes, in electric furnaces.

High-quality steels, on the contrary, are generally made in electric furnaces; only in special cases can they be melted by the open-hearth process.

The letter A is appended to the specification number designating the grade of a high-quality steel in those cases when the same grade may be classed either as a quality or as a high-grade steel.

2. Carbon Steels and Their Characteristics

Structural carbon steels. Structural carbon steels are divided into three groups according to quality: *ordinary quality*, *higher quality* and *quality* steels.

The chemical compositions and properties of *structural carbon steels* of ordinary and higher quality types are indicated in the State Standards. According to these Standards all structural carbon steels are divided into three groups:

a) In the first group are steels with guaranteed mechanical properties; their chemical composition is optional (not obligatory), except for sulphur and phosphorus; the main grades of this group are designated as follows: БСт. 3, БСт. 4, БСт. 5, БСт. 6 (Bessemer steels), and Ст. 1, Ст. 2, Ст. 3, Ст. 4, Ст. 5, Ст. 6, Ст. 7 (open-hearth steels).

b) Steels of the second group have guaranteed chemical composition; the main grades of this group are designated as follows: Б09, Б16, Б23, Б33 (Bessemer steels), and М09 кп, М12 кп, М18 кп, М18, М21, М26, М31, М44, М56 (open-hearth steels); the numbers indicate the average carbon content in "points", or hundredths of 1 per cent, and the letter symbols "кп" designate rimmed steels (specification numbers without "кп" indicate killed steels).

c) Steels of the third group (higher quality steels) have guaranteed compositions and mechanical properties and are designated as follows: M09, M12, M16, M18a, M21a, M26a, M31a, M44a, M56a.

The higher the number designating the grade of carbon steel of ordinary quality belonging to the first group, the higher is its carbon content; thus, the carbon content in steel of grade Ст. 1 is not above 0.12 per cent, in steel of grade Ст. 4 up to 0.25 per cent. The figures 1, 2, 3, etc., in standard designations of these grades indicate only the conventional number of steel. For example, the figure 3 in designation of grade Ст. 3 does not at all signify, as is sometimes wrongly believed, that this steel contains 0.3 per cent carbon: the actual carbon content in steel of this grade is not over 0.22 per cent.

The greater the proportion of carbon in steel, the higher are the values of its tensile strength, yield point and hardness, and the lower those of elongation and resilience (impact strength). In other words, the greater the carbon content in steel, the higher, all other things being equal, are its strength, elasticity (determined by yield point) and hardness, and the lower its ductility (characterised by elongation) and impact strength.

The technological and mechanical properties of these steels depend largely upon their carbon content. Cold workability and weldability rapidly deteriorate with an increase in carbon. Steels of grades M09 кп-M18 кп usually are used to manufacture parts by cold working methods, whereas the steel of grade M18 is used for welded structures. An increase in carbon content also rapidly impairs the flame cuttability of the steel. Only the grades Ст. 1-Ст. 4 and M09 кп-M31 can easily be flame-cut. On the other hand, the ability of steel to harden, i. e., its liability to increase in hardness through hardening treatment, improves greatly with an increase in carbon content.

The uses of steels of the first and second groups are determined basically by the following considerations. Steels of the second group are generally used for welded structures because of their guaranteed chemical composition, since the latter (or, more precisely, the carbon content) largely determines the weldability. Therefore, the main part of welded structures is made of steel M18, but not of grade Ст. 3. Parts undergoing the strengthening heat-treatment (normalising, and hardening with subsequent tempering) are also, in general, manufactured of steels of the second group, since to have certain mechanical properties steel must have a certain carbon content. Parts which must possess certain mechanical properties and which are manufactured of rolled steel without subsequent heat-treatment, i. e., parts with mechanical properties determined by the mechanical properties of rolled stock, are usually made of steel of the first group.

Structural carbon steels of ordinary quality are generally used for making hot-rolled, so-called ordinary stock: plates, sheets, rods, channels, beams, angles, tubes, wire bars, and light-weight general-purpose forgings. Steels of ordinary quality are not used for manufacture of cold-drawn (sized) steel rods and wire.

The chemical composition and mechanical properties of carbon quality structural steels are specified by the U.S.S.R. State Standards. Quality structural carbon steels are divided into two groups according to their manganese contents: the first group with normal manganese content (not exceeding 0.80 per cent), and the second one higher in manganese (minimum manganese content 0.70 per cent).

Steels of the first group are designated by two-digit numbers indicating the average carbon content in "points" or hundredths of 1 per cent as: 05 кп, 05, 08 кп, 08, 10 кп, 10, 15 кп, 15, 20 кп, 20, 25 . . . 85 (23 grades in all). Grades 05 кп, 08 кп, 10 кп, 15 кп and 20 кп are known as rimmed steel (deoxidised only by manganese), and other grades as killed steel (deoxidised by silicon and manganese). Rimmed steel is characterised by its very high ductility and is especially suitable for deep drawing operations.

Steels of the second group (17 grades in all) are higher in manganese content and are designated by two-digit numbers indicating the average carbon content in hundredths of 1 per cent, and by the letter symbol "Г" indicating manganese: 20Г, 65Г, etc. In the designation of eight of the grades of the second group, the figure 2 is annexed to the letter symbol (10Г2, 30Г2, etc.); the average manganese content in steels of these grades is 1.6 per cent.

Quality structural carbon steels are characterised according to quality specifications as follows:

1) the mechanical properties of these steels are somewhat higher than those of ordinary steels (Table 3);

2) quality steels are specified either by chemical composition or by mechanical properties.

Quality structural carbon steels are used for plates (both hot-rolled and cold-rolled), rods and tubes (hot-drawn and cold-drawn), wire, and special-purpose forgings.

Structural carbon steels also include the so-called *free-cutting steels* that are used for small-size machine parts (in particular for various fastenings, studs and screws), especially for those produced by automatic machining. To be suitable for high-speed machining operations, these steels should possess high *machinability*, i. e., free machining properties. The poor machinability of plain low-carbon steels (not free-cutting steels) is caused by their high ductility and toughness. When ordinary steels are machined, the chip comes away in the form of a long continuous helix, which causes friction and consequent intensive heating and wearing of

Table 3

Comparison Between the Specified Mechanical Properties of Ordinary Quality and Quality Structural Carbon Steels
(steels of approximately the same carbon content are compared)

Grades to be compared		Tensile strength		Yield point, not less than		Elongation, not less than	
Ordinary	Quality	Ordinary	Quality	Ordinary	Quality	Ordinary	Quality
Ст. 1	10	32-40	36-45	—	22	33	32
Ст. 3	20	40-50	44-54	25	26	25	26
Ст. 5	30	50-62	50-62	29	30	19	22
Ст. 6	45	60-72	64-76	32	36	14	17
Ст. 7	55	⩾70	71-83	35	40	10	13

the cutting tool. Therefore when ordinary steels are machined, reduced cutting speed and frequent change of tools are necessary, and these decrease the productivity of any machine tool, especially an automatic one.

To increase the machinability of low-carbon steels it is necessary to reduce their ductility. In this case there will not be a long continuous helical chip but a breaking off (loose) chip, and the friction between this and the cutting tool will be greatly reduced. The addition of considerable amounts of sulphur and phosphorus (up to 0.30 and 0.15 per cent respectively) in free-cutting steels reduces the ductility and thus improves the machinability of low-carbon steels. The U.S.S.R. State Standard specifies four grades of free-cutting steel: A12, A20, A30 and A40Г. In spite of the increased content of such harmful elements as sulphur and phosphorus, the mechanical properties of free-cutting steels are not altogether bad: the tensile strength and hardness of these steels are somewhat higher, elongation slightly less, and only the reduction of area considerably less (by $1^1/_2$ to 2 times) than those of ordinary and even quality steels.

This is explained as follows. Sulphur, if it is contained in the form of FeS distributed in the metal structure as long inclusions at the grain boundaries, has a very injurious effect upon the properties of the steel. But when sulphur is contained in the steel structure in the form of double sulphide MnS-FeS as small globules the harmful effect is considerably reduced. In free-cutting steels the manganese content is intentionally increased in order to ensure the forming of these globules of double sulphide (MnS-FeS).

Carbon Tool Steels. The chemical composition of *carbon tool* steels is specified by the U.S.S.R. State Standards. There are

available eight grades of quality tool steel: У7, У8, У8Г, У9, У10, У11, У12, У13 and eight grades of high-quality tool steel: У7А, У8А, У8ГА, У9А, У10А, У11А, У12А and У13А. The letter symbol У designates carbon steel; the figures annexed to letter symbol У indicate the carbon content in tenths of 1 per cent. The letter symbol Г, designating manganese, indicates higher manganese content; the symbol А at the end of a grade designation number indicates, as usual, that the steel belongs to the high-quality class.

The uses of tool carbon steels are determined by their carbon contents. The higher the carbon content, the greater the hardness of steel in as-hardened condition and the lower its impact strength. Therefore grades У7 and У7А are used for fitter's tools, which are subjected to shocks during use (point-tools, chisels, marking irons, hammers, sledge hammers, etc.). Tools, which are subjected to shocks in service and at the same time require higher hardness, such as matrices, dies, punches, shear blades for metal cutting, pneumatic chisels, etc., are made from steel of grades У8, У8А, or less often of grades У8Г, У8ГА, У9, У9А.

Grades У10, У10А, У11, У11А, У12 and У12А are used primarily for such cutting tools as drills, milling cutters, breaches, reamers, taps, files, etc., and for various gauges. Grades У13 and У13А are used for tools, which are not subjected to shocks and should possess very great hardness, such as chisels for cutting files, scrapers, drawing dies, boring tools, etc.

3. Alloy Steels and Their Characteristics

For most of grades of alloyed steels the following standard system of designations is adopted in the U.S.S.R.

Grades of *alloy structural* steels are designated by two-digit numbers indicating the carbon content in *hundredths* of 1 per cent, and by one or more capital letters indicating the alloying elements contained in the given steel. These letter symbols have the following meaning:

Х — chromium	В — tungsten	Т — titanium
Н — nickel	Ф — vanadium	Б — niobium
Г — manganese	М — molybdenum	Ю — aluminium
С — silicon	К — cobalt	

If the content of alloying element in steel is equal to or less than 1 per cent, the corresponding letter symbol is usually not followed by a number, but if the alloying element is above 1 per cent, the letter symbol is followed by a figure indicating the *approximate* content of this element in per cent. However, it is necessary to be rather cautious as to the latter: in some cases the real content of alloying

element differs from that indicated in the grade designation number. For instance, in steel of grade 40HM the nickel content is 1.50-2.00 per cent, and in grade 20X2H4 it ranges from 3.25 to 3.75 per cent.

Grades of *alloy tool* steel are designated by one-digit numbers (i. e., by one figure) indicating the carbon content in *tenths* of per cent, and by one or more letter symbols indicating the alloying elements contained. The meaning of the letter symbols is the same as for the alloy structural steels. If the carbon content in alloy tool steel is 1 per cent or more, the numbers are not put before the letter symbols. For example, the carbon content in steel of grade ХВГ ranges from 0.90 to 1.05 per cent.

The same index system is used for designation of *high-alloy structural* steels with special physico-chemical properties; but in the majority of cases the numerals indicating carbon content are not put before the letter designation. The figures (one-digit) are used only in the designations of those grades of steel that differ from each other in carbon content, but not in alloying elements. For instance, four grades of stainless steel alloyed with equal proportions of chromium (12 to 14 per cent) are designated as follows: 1X13 (carbon content under 0.15 per cent), 2X13 (carbon content from 0.16 to 0.24 per cent), 3X13 (carbon from 0.25 to 0.34 per cent), and 4X13 (carbon from 0.35 to 0.45 per cent).

Apart from these simple and rational specifications, there are other designations for some groups of steel. Thus, two standard grades of *high-speed* steel are designated by the letter symbol "P" with a subsequent numeral (9 or 18) indicating content of tungsten, which is the predominant alloying element in this kind of steel: e. g., P9 and P18.

Grades of steel used for *ball bearings* are designated by letter symbols ШХ (derived from the Russian words for "ball bearing chromium steel"), followed by figures indicating the approximate chromium content in tenths of per cent, e. g., ШХ6 (chromium content from 0.4 to 0.7 per cent), ШХ9 (chromium from 0.9 to 1.2 per cent), ШХ15 (chromium from 1.3 to 1.65 per cent).

Grades of steel for *electrical purposes* (transformer and dynamic irons) are designated by the letter symbol Э and two-digit or three-digit numbers: the first digit represents the approximate silicon content in per cent; the second digit, the conventional designation of guaranteed magnetic properties; the third digit (which can be only zero) indicates that a given steel belongs to the cold-rolled class. The absence of the third digit indicates that the steel has been hot-rolled. For instance, grade Э41 is hot-rolled electrical steel containing about 4 per cent silicon (transformer iron) with normal specific core losses when remagnetised at 50 cycles.

As compared with carbon structural steels, alloy structural steels are characterised by higher tensile strength and yield point (Table 4). The latter property is of prime importance, since the higher values of yield strength indicate higher values of permissible (rated) stresses.

Table 4

Comparison Between the Mechanical Properties of Carbon and Alloy Structural Steels in Heat-Treated Condition
(hardening and high tempering)

Grade of steel	Tensile strength	Yield point	Elongation	Contraction	Impact value
	not less than				
40	60	34	19	45	9
45	64	36	17	40	8
40X	80	65	12	40	7
40XC	90	70	12	45	6
40XH	83	65	13	45	8
OXH3M (0.35%C)	100	84	13	45	8
OXH4M (0.35%C)	105	90	12	40	8

Among the structural steels is especially singled out the group of steels used for springs; the chemical composition and mechanical properties of these steels are specified by the U.S.S.R. State Standard. Carbon spring steels are characterised by high carbon content ranging from 0.60 to 0.90 per cent, whereas in alloy spring steels the carbon content ranges from 0.45 to 0.70 per cent.

High-speed steels hold a unique place among alloy tool steels. The remarkable property of these steels is their high *mechanical strength* at elevated temperatures (heat resistance). By this property is understood the power of hardened steel to retain its hardness when heated to elevated temperatures. The heating of a cutting tool during use causes self-tempering of the steel with a subsequent loss of hardness. Tools of hardened carbon steels usually retain their hardness up to 200°C. When heated above this temperature they begin to soften rapidly, losing their cutting properties and becoming "blunted", so that the further machining of metal becomes impossible.

The high-temperature strength of alloy tool steels is somewhat higher. High-speed steels retain their hardness even at temperatures of 600°C. A temperature of 600°C is that of low red heat, so the high heat resistance of high-speed steels is usually called a "red-hardness". High-speed steels are alloyed with tungsten, chromium and vanadium. The U.S.S.R. State Standard specifies two grades of high-speed steels—P18 (containing 18 per cent tungsten), and P9 (containing 9 per cent tungsten).

High-alloy steels *with special physico-chemical properties* include: stainless, acid-resisting, non-scaling, heat-resisting steels, steels with high ohmic resistance, electrical, magnetic, non-magnetic, and wear-resisting steels.

Stainless steels possess a high *resistance to* atmospheric *corrosion* and to attack by humidity at ordinary and elevated temperatures. All types of stainless steel are alloyed with chromium. The main grades of stainless steel are as follows: 1X13, 2X13, 3X13 and 4X13. The stainless steels are used for steam turbine blades, valves for hydraulic pumps, household articles and utensils (knives, forks, washing-machine tanks, etc.).

Acid-resisting steels possess high *corrosion resistance* to attacks by various acids and other chemicals. Like stainless steels, all acid-resisting steels are alloyed with chromium. Most acid-resisting steels also contain nickel, and to some of them are added manganese, molybdenum, titanium and niobium. Titanium and niobium are added specifically to improve the weldability of these steels. When acid-resisting steels, which do not contain titanium or niobium are welded, carbides form, on account of which areas of metal near to the weld seam lose to a considerable extent their corrosion resistance, so that they rapidly corrode in service. But when acid-resisting steels, containing titanium or niobium, are welded, no structural transformations occur in the weld area, which retains its high corrosion resistance. The most widely used acid-resisting steels are: X17, X25, X28, 1X18H9T, 2X18H9, X18H11Б, X18H12M3T.

Non-scaling steels (which are sometimes wrongly called heat-resisting) have high resistance to gas corrosion at elevated temperatures. All non-scaling steels contain chromium, and some are also alloyed with silicon, aluminium, nickel and titanium. Steels of grades X18H25C2 and X25H20C2 are not liable to scaling when heated to about 1,000 to 1,100°C. The non-scaling steels are used for parts of furnace conveyors, muffles, thermocouple protective tubes, etc.

Non-scaling steels that retain much of their strength at elevated temperatures are called *heat-resisting* steels. All heat-resisting steels are alloyed with chromium, and they may also contain molybdenum (X5M), molybdenum and silicon (X6CM, X7CM, X10C2M), nickel and silicon (X13H7C2), nickel, tungsten and molybdenum (4X14H14B2M, 1X14H14B2M), nickel, molybdenum and titanium (X18H12M3T), nickel, silicon, tungsten and molybdenum (X14H14CB2M). The acid-resisting chromium-nickel steel 1X18H9T alloyed with titanium is also classed as a heat-resisting steel. The less alloyed steels, in particular the so-called "silchroms" (steels, alloyed with chromium, silicon and molybdenum, of grades X6CM, X7CM, X10C2M), are used for parts employed at about 350 to 500°C. They are also used for valves of car engines. The higher

alloyed steels are employed at higher temperatures (500 to 650°C). These steels are suitable for parts of gas turbines and jet engines. The nickel- or cobalt-base alloys are used at even higher temperatures (650 to 800°C).

Steels of *high ohmic resistance* possess high electrical resistivity, attaining 1.0 to 1.3 ohm $\times$ mm^2 $\times$ m, i. e., 10 to 12 times that of iron. At the same time they are non-scaling. Therefore these steels are used for wire and ribbons, which are employed for electric furnace heaters. All these steels are alloyed both with chromium and aluminium. They are also characterised by very low carbon content (less than 0.1 per cent). The less the carbon content in such steels, the higher the temperature they can withstand for long period in service.

Electrical steels are divided into two groups: *dynamic* and *transformer* irons. Electrical steels are alloyed with silicon, whose content in dynamic irons of grades Э1, Э2 and Э3 is approximately equal to 1, 2 and 3 per cent respectively, and in transformer iron of grade Э4, about 4 per cent (the U.S.S.R. State Standard does not specify the precise chemical composition of these steels, since grades of electrical steels are distinguished according to their magnetic properties). In electrical steels the carbon content should not exceed 0.1 per cent, since it greatly impairs the magnetic properties of these steels. Electrical steels are characterised by high magnetic induction, extremely low coercive force and low watt losses (i. e., low heat losses by electric power required for remagnetising). Dynamic steel sheets are used for stamping stock for poles and other magnetic conductors of electrical machines, and sheets of transformer iron are used for transformer cores.

Unlike the electrical steels, those used for *permanent magnets* should, in addition to high magnetic permeability, possess high coercive force. Steels for permanent magnets are alloyed with chromium, tungsten and cobalt. The U.S.S.R. State Standard specifies five grades of steels for permanent magnets, two chromium (EX and EX3), one tungsten (Е7В6), one chromium-cobalt (EX5К5), and one chromium-cobalt-molybdenum (EX9К15М). The cobalt steels possess especially high magnetic properties. Recently magnet steels have been increasingly replaced by cast iron-nickel-aluminium alloys of types such as Alni, Alnisi, Alnico and Magnico.

Non-magnet (more precisely, *low-magnet*) steels are used for those parts of electrical machines that must be as strong as steel and as non-magnetic as copper or bronze, e.g., bandage rings for turbogenerators. Among the non-magnet steels, grade Н9Г9, containing 0.50 per cent carbon, is most widely used.

Wear-resisting steels are especially advantageous because of their resistance to wear, i. e., resistance to abrasion by sand, earth, rock, ore and other hard substances, which is accompanied by cold harden-

ing of steel. Wear-resisting steels are used for jaw plates and other parts of rock crushing machines, cases of edge and ball mills, dipper teeth for power shovels and excavators, railroad crossings, frogs, switch points, guard rails, etc. Among the wear-resisting steels only one grade may be mentioned, i. e., high-manganese steel, occasionally referred to as Hadfield steel (after the name of prominent English metallographer R. A. Hadfield who first developed it). This steel contains 1.0 to 1.3 per cent carbon and 11 to 13 per cent manganese. Steel of this grade is usually designated as Г13.

4. The Iron-Carbon Phase Diagram

The diagram in Fig. 1 gives the iron-carbon alloy system. This diagram refers to conditions of iron-carbon alloys, when, in the course of their primary and secondary crystallisations, carbon separates from liquid and solid solutions not in the pure form (as a graphite), but as a chemical compound, i. e., as an iron carbide or cementite (Fe_3C). Consequently this constitutional diagram (or phase diagram) covers only steels and white irons. For the study of gray irons this diagram alone is insufficient.

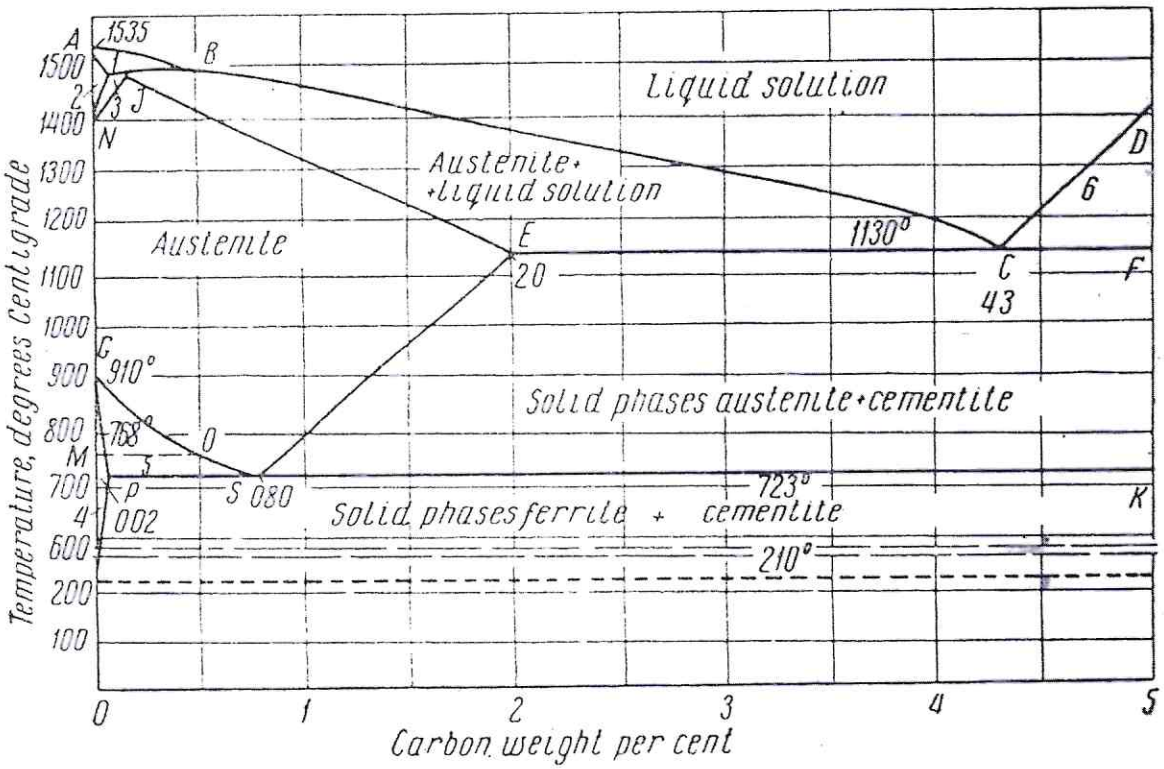

Fig. 1. Iron-carbon phase diagram.

1 — liquid solution + solid solution; *2* — solid solution; *3* — solid solution + austenite; *4* — solid solution (ferrite); *5* — ferrite + austenite; *6* — liquid solution + + cementite.

In several areas of the diagram there are indicated names of the phases existing at temperatures and concentrations determined by the boundaries (lines) of these areas.

It should be noted that by the term "*a phase*" is meant a *perfectly homogeneous* (*in physical and chemical respects*) *portion of an alloy* (*with one common pattern of space lattice*) *separated from the remainder by distinct bounding surfaces*. A liquid solution and a solid solution of the same chemical compositions constitute different phases, because their respective physical, or aggregate states are different. The two allotropes of iron, alpha iron and gamma iron, are also different phases because their space lattices differ.

The portions of the diagram which lie between the *liquidus* and *solidus* lines (*ABCD* and *AIECF* respectively) represent the process of *primary* crystallisation on freezing the alloys, i. e., the solidification of the liquid solution, whereas those areas between lines *GSECF* and *PSK* represent the process of *secondary* crystallisation, i. e., decomposition of solid solution or austenite.

Most of the heat-treating operations applied to steel (some kinds of annealing, normalising, hardening, carburising, etc.) are connected with the processes of secondary crystallisation. We shall therefore consider more closely the left-hand bottom portion of the iron-carbon phase diagram (Fig. 2).

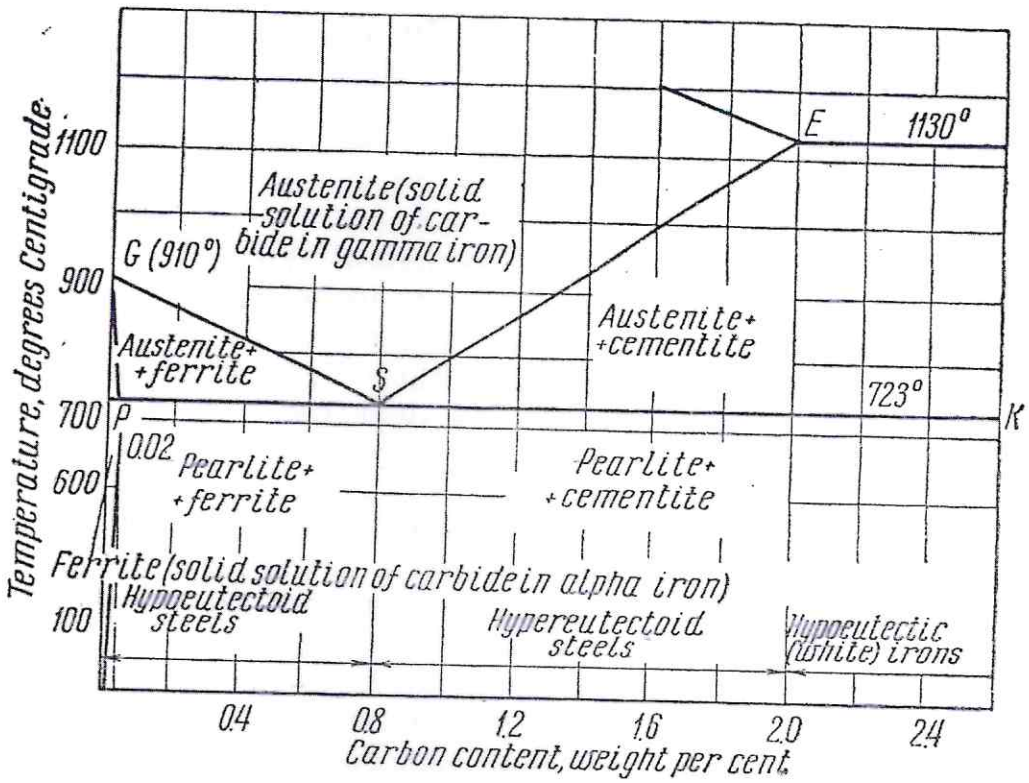

Fig. 2. Left-hand bottom ("steel") portion of the iron-carbon phase diagram area of the secondary crystallisation of steel).

At temperatures above lines *GS* and *SE* only one phase, *austenite*, exists.

Each point on the lines *GS* and *SE* represents temperature, at which the process of austenite decomposition begins in freezing steel. In hypoeutectoid steels containing up to 0.8 per cent carbon, decomposition of austenite takes the form of the separation from it of a new phase, called *ferrite*. Separation of ferrite takes place continuously during the decomposition of cooled austenite, beginning at temperature A_3 (on line *GS*) and reaching completion at temperature A_1 (on line *PSK*). As the ferrite separates, the carbon concentration in the austenite progressively increases, reaching the value of 0.8 per cent (the *S* point) at a temperature of 723°C (line *PSK*).

In hypereutectoid steels containing more than 0.8 per cent of carbon the decomposition of austenite takes the form of the separation from it of a *cementite*. The separation of cementite takes place continuously during the decomposition of cooled austenite, starting at a temperature A_3 (or the A_{cm} point on line *ES*) and being completed at a temperature A_1 (on the *PSK* line). As the cementite separates, the carbon concentration in the austenite continuously decreases, also reaching the value of 0.8 per cent (the *S* point) at a temperature A_1 (on line *PSK*).

Thus, in slow cooling any steel, hypoeutectoid as well as hypereutectoid, has in its structure austenitic grains containing 0.8 per cent carbon as well as ferrite or cementite. At the A_1 point, the process of secondary crystallisation ends with the *eutectoid decomposition* of the remaining austenitic grains. The austenite becomes a conglomerate consisting of two phases, ferrite and cementite.

The structure of steel is formed as a result of these transformations. However, the knowledge of the iron-carbon phase diagram only is insufficient to predetermine precisely what this structure will actually be. This will be explained by the following example.

As already indicated, at the A_1 temperature the secondary crystallisation of steel ends with the eutectoid decomposition of austenite into a conglomerate consisting of two phases, ferrite and cementite.

On relatively slow cooling, this conglomerate of ferrite and cementite becomes sufficiently distinctive for its structure to be well enough examined at a magnification of about 500 diameters. Such ferrite-cementite mixture is known as *pearlite*. With more rapid cooling the austenite is also transformed into a mixture of ferrite and cementite, but the structure of this conglomerate is less distinctive, so that in order to examine it substantially greater magnifications (about 1,000 to 1,500 diameters) must be used. Such a mechanical mixture of ferrite and cementite is called *sorbite*. With

still more rapid cooling rates a ferrite-cementite conglomerate will also be obtained, but the structure of this modification of pearlite is so fine that it can be distinguished only by means of an electron microscope. Such an aggregate of ferrite and cementite is called *troostite*.

Thus, in all three cases the phase composition of steel is of exactly the same pattern, i. e., ferrite and cementite, but in each case the structure of the steel and its properties will vary markedly.

From this example an extremely important conclusion may be drawn, i. e., that the *formation of the structure of steel* (or of any other alloy) *is determined by the kinetics of phase transformations*, i. e., by the process of this transformation with time. The practice of heat-treating steel is based to a considerable extent, if not entirely, on a consideration of the kinetics of phase transformations that take place during the secondary crystallisation of steel.

It is also very important to note that under real conditions the transformation temperatures A_1 and A_3 differ from those indicated in the phase diagram. Under real conditions the austenite transforms into a ferrite-cementite conglomerate not at 723°C, as is indicated in the diagram, but at a slightly *lower* temperature. The reverse change of the ferrite-cementite aggregate to austenite on heating takes place at temperatures higher than 723°C, i.e.,

$$A_1 \geqslant 723^\circ \text{ on heating,}$$

$$\text{and } A_1 \leqslant 723^\circ \text{ on cooling.}$$

The real value of the A_1 temperature depends upon the speed of heating or cooling.

Thus, it is evident that, firstly, the iron-carbon phase diagram cannot explain the process of formation of a specific steel structure under real conditions of heat-treatment, and secondly, the real temperatures at which phase transformations occur under the same conditions differ in a greater or lesser degree from those indicated in the phase diagram. The question naturally arises as to what the iron-carbon phase diagram can indicate in such a case, and what help it can afford in the practical accomplishment of heat-treating operations.

The answer is as follows: firstly, on the basis of this diagram it can be quite accurately ascertained what phase composition will have an alloy in a condition of equilibrium at any temperature, and consequently, what phase composition is likely to arise in an alloy being in a state of non-equilibrium. The equilibrium condition of a system determines the direction of spontaneous processes that take place or may take place in a system which is not in a state of equilibrium.

Secondly, a sufficient approximation of an alloy to a condition of phase equilibrium is reached on its being slowly heated to high temperatures. So the actual temperatures of phase transformations closely coincide with those indicated by the diagram. Therefore the selection of the temperatures of heating for annealing, normalising and hardening is made on the basis of the iron-carbon phase diagram.

Thirdly, this diagram facilitates the establishing of the temperature ranges of phase transformations.

A condition of steels approaching equilibrium is attained on heating very slowly from temperatures above the A_3 point and is accomplished practically by annealing. Depending upon the carbon content, the structure of annealed carbon steels at room temperatures will be composed of:

pearlite and ferrite: hypoeutectoid steels (less than 0.8 per cent carbon);
pearlite: eutectoid steel (0.8 per cent carbon);
pearlite and cementite: hypereutectoid steels (more than 0.8 per cent carbon).

As to terminology, there must be made one reservation. It is often said that "the structure of steel is composed of pearlite and ferrite". This is not altogether correct: each steel, as well as each crystalline body, consists of grains (crystals). Thus, it is more exact to say that "the structure of steel is composed of pearlite and ferrite grains".

However, if it is stipulated that by ferrite is meant not only a certain phase but an aggregate of all ferrite grains, and under the term pearlite not only the structural constituent but an aggregate of all pearlite grains, the above definition may be further abbreviated in the following manner: "The structure of steel consists of pearlite and ferrite."

5. Phase Transformations in Alloy Steels

The transformations occurring under conditions of equilibrium in steels alloyed with one element only should be studied by means of ternary diagrams, i. e., the iron-carbon-alloying element, while those in steels alloyed with two elements should be investigated by means of quarternary diagrams, etc. But the ternary diagrams are very complicated and not yet sufficiently studied, so that their practical usage involves considerable difficulties. Still more complex are the quarternary diagrams. Therefore in the study of alloy steels binary phase diagrams of the iron-alloying element type are usually considered; in this way is studied the influence of alloying elements

on the positions of lines and transformation points in the iron-carbon constitutional diagram.

With regard to chief alloying elements, all phase diagrams of type iron-alloying elements may be reduced to two main types (Fig. 3). They differ from each other mainly in the manner in which

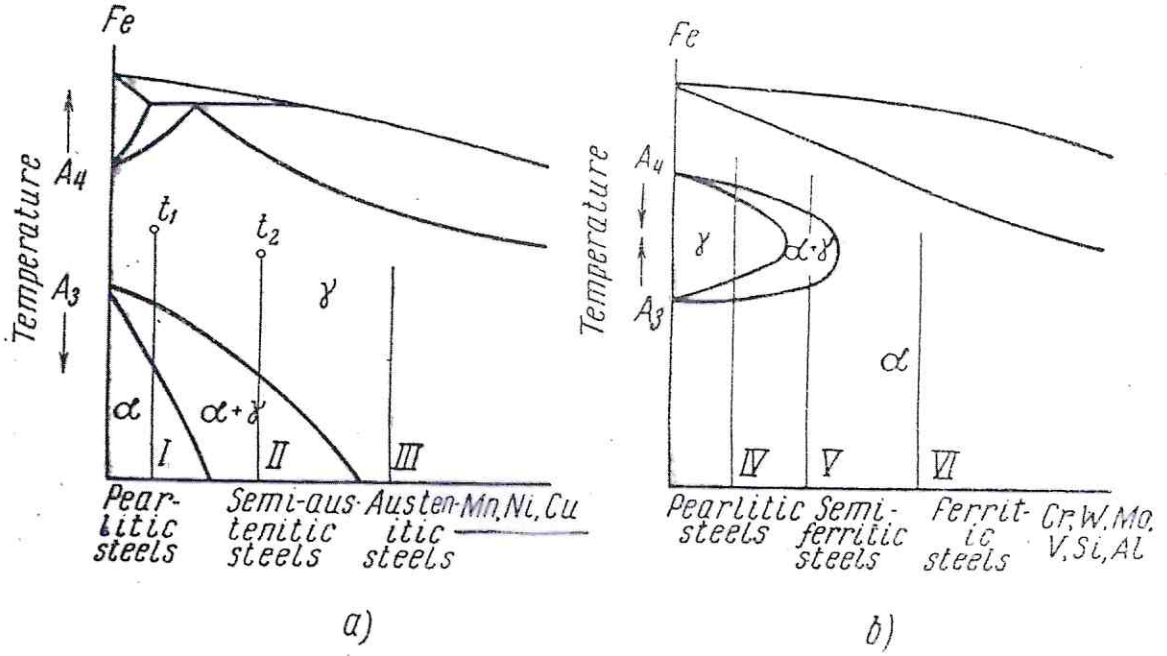

Fig. 3. Patterns of the main phase diagrams of the iron-alloying element type.

the alloying elements influence the widening or narrowing of the area where the gamma solution, i.e., solid solution of alloying element in gamma iron, exists.

Some alloying elements (Ni, Mn) broaden the area of gamma iron (Fig. 3, *a*) and hence depress the A_3 point and raise the A_4 point. The greater the content of alloying element in an alloy, the higher the temperature at which the transformation A_4 takes place and the lower the temperature at which the transformation A_3 occurs. From this it follows that alloys low in alloying element (alloy I) consist (at room temperature) of a solid solution of alloying element in alpha iron, called an alpha solution. As the temperature is raised complete phase recrystallisation occurs, and at a temperature t_1 the alloy I consists of gamma solution.

At room temperature alloys with larger concentrations of alloying element (alloy II) are made up of two phases, alpha solution and gamma solution. As the temperature rises the alpha solution changes to gamma solution, and at the temperature t_2 the alloy II consists of the gamma solution only. Thus, in alloys like alloy II only partial phase recrystallisation takes place.

Finally, with a still greater content of alloying element (alloy III) the alloy consists of gamma solution only at all temperatures. Consequently in alloys of type III the phase recrystallisation does not occur at all.

The presence of carbon in these alloys, turning them into alloyed steels, produces in general no effect whatsoever upon the character of the phase transformation process described, though it gives rise to some quantitative changes.

Low alloy steels (like alloys of type I) at room temperature are made up of two phases—ferrite (alpha solution) and carbide (cementite if the alloying element does not form carbides, or complex carbide of type Fe_mMe_nC if the alloying element is carbide forming). According to the carbon content, these steels in annealed condition at room temperature may be classed as hypoeutectoid (ferrite and pearlite), eutectoid (pearlite) and hypereutectoid (pearlite and secondary carbides). Complete phase recrystallisation, similar to that taking place in carbon steels, occurs in these steels at temperatures A_1-A_3 (Fig. 1). These steels are known as *pearlitic*.

At room temperature the structure of the higher alloyed steels (corresponding to alloys of type II) in an annealed condition is composed of austenite and pearlite (plus ferrite and secondary carbides, depending upon the carbon content). At a temperature range A_1-A_3 (Fig. 1) partial phase recrystallisation occurs: pearlite + ferrite, pearlite, pearlite + secondary carbides transform into austenite. The austenite phase, which existed in structure at room temperature, does not take part in the phase recrystallisation process. These steels are known as *semi-austenitic*.

Highly alloyed steels (corresponding to alloys of type III) have the austenitic structure at all temperatures. The phase recrystallisation does not take place in these steels. They are classed as the austenitic steels.

Other alloying elements (chromium, tungsten, molybdenum, vanadium, silicon, aluminium) tend to narrow the area in which the gamma solution exists (Fig. 3, *b*). At room temperature all alloys of iron with the alloying elements of this group are composed of alpha solution only. Alloys IV (with low content of alloying element), V (with higher content of alloying element) and VI (with high content of alloying element) differ from each other in their tendency to phase recrystallisation, i. e., alloys IV are capable of complete phase recrystallisation, in alloys V the phase recrystallisation covers only part of the alpha solution, and in alloys VI it does not occur at all.

Steels slightly alloyed with elements of this group (corresponding to alloys IV) in annealed condition and at room temperature are composed of pearlitic grains and, according to carbon content, of

ferritic grains and secondary carbides. These steels are capable of complete phase recrystallisation. They are known as pearlitic steels.

Only partial recrystallisation can occur in steels with a higher content of alloying elements (corresponding to alloys V). These steels are classed as *semi-ferritic*.

Finally, in highly alloyed steels (corresponding to alloys VI) phase transformations do not take place at all. At all temperatures their structure is composed of alpha solution or, if the steel is alloyed with a carbide-forming element and the amount of carbon present is substantial, of alpha solution and carbides. These steels are classed as *ferritic*. The ledeburite eutectic is found in the structure of high-carbon steels alloyed both with the elements of the first and second groups. As the result of forging the ledeburite eutectic is broken down into a series of isolated ledeburite carbides. These steels are identified as steels belonging to the carbide class.

Thus, alloy steels may be divided into six structural classes in accordance with their ability to undergo phase recrystallisation and their phase composition at room temperature. If the practically irrelevant semi-austenitic and semi-ferritic classes are excluded, most alloy steels may be divided into four main structural classes — pearlitic, austenitic, ferritic and carbide.

All low-alloy structural steels, the greater part of alloy tool steels (with a relatively low carbon content), and stainless steels of grades 1X13-4X13 are *pearlitic*. The structure of a steel of this pearlitic class in annealed condition is represented by Fig. 4.

The highly and complexly alloyed acid-resisting steels of grades 0X18H9, 1X18H9, 1X18H9T, 2X18H9 and X13H4Г9, non-magnetic steels and wear-resisting steel of grade Г13 are classed as *austenitic*. The structure of the steel of austenitic class is shown in Fig. 5. The fact that the most typical acid-resisting steels are specified as austenitic is explained by the following considerations; as is well known from the theory of corrosion, alloys with a homogeneous (single-phase) structure are more resistant to corrosion than those with a heterogeneous structure. It is clear, too, why non-magnetic steels should have the austenitic structure, since among all the phases of iron-carbon alloys only austenite is non-magnetic. The high wear-resisting properties of articles made of high-manganese steel Г13 are explained as follows: the austenite transforms into martensite by abrasive grinding accompanied by cold hardening, so that the surface hardness of the steel increases greatly.

The *ferritic* class includes the highly alloyed chromium non-scaling steels of grades X25, X25T, X28, high-ohmic resistance steels, and transformer irons. The structure of steels of the ferritic class is shown in Fig. 6.

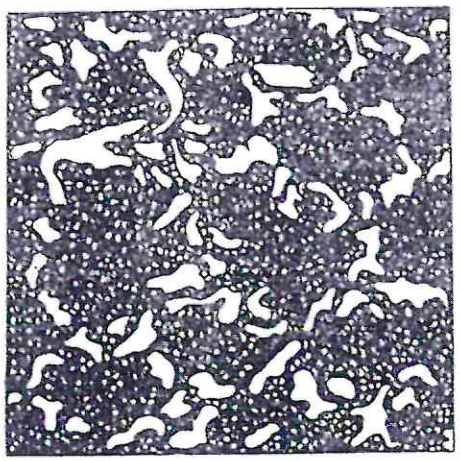

Fig. 4. Structure of alloyed steel of the pearlitic class in annealed condition. Ferrite and sorbite, 300 ×.

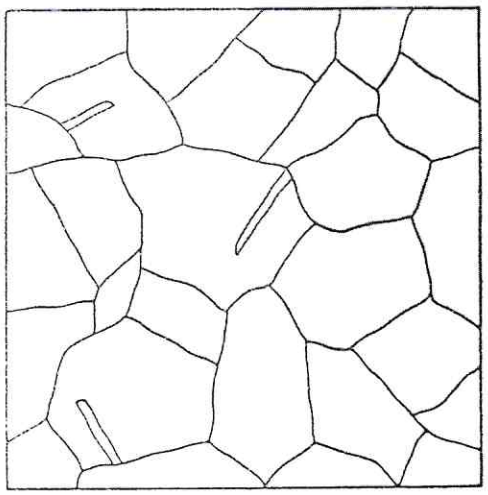

Fig. 5. Structure of alloyed steel of the austenitic class. Austenite, 300 ×.

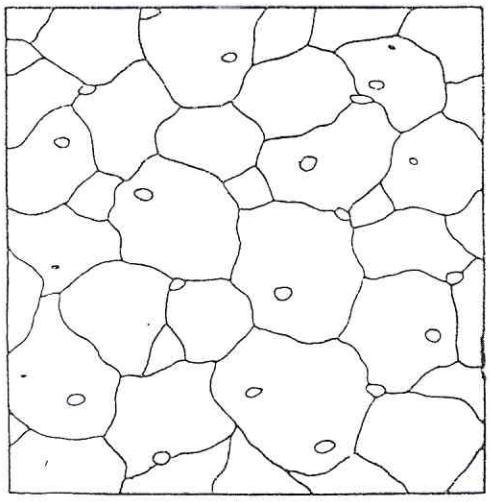

Fig. 6. Structure of alloyed steel of the ferritic class. Ferrite and carbide, 300 ×.

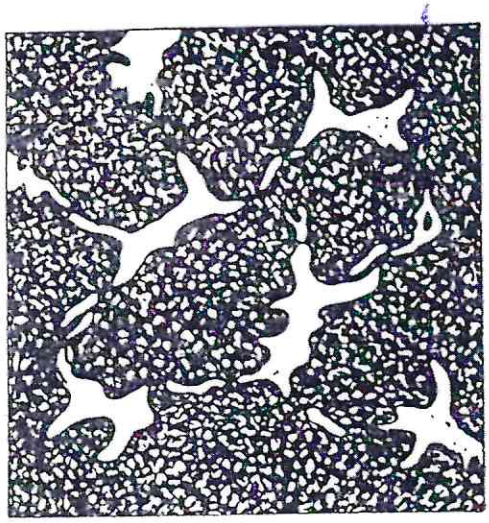

Fig. 7. Structure of high-speed steel in as-cast condition, 500 ×.

The *carbide* class comprises some tool steels, such as chromium steels of grades X12 and X12M, and high-speed steels of grades P9 and P18. The structure of steels of the carbide class in as-cast condition contains the ledeburite eutectic (Fig. 7). Forging and rolling cause the decomposition of the ledeburite eutectic into separate grains, the structure of forged steels of the carbide class in annealed condition being composed of grained pearlite and of larger grains of secondary and ledeburite carbides (Fig. 8).

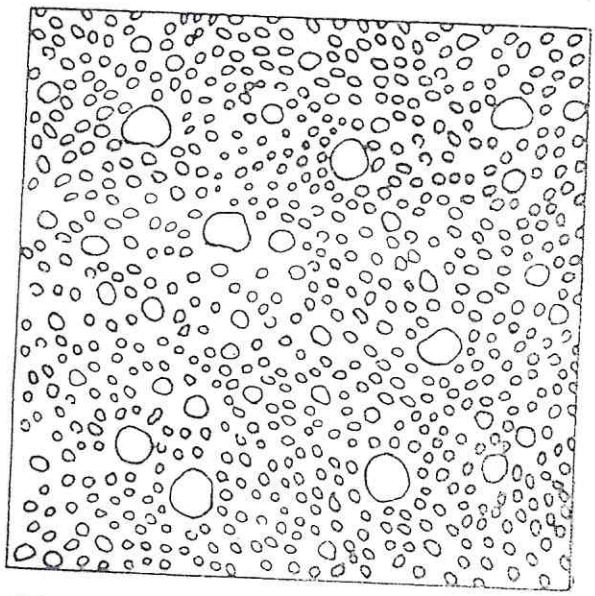

Fig. 8. Structure of alloyed steel of the carbide class (high-speed steel of grade P18) in annealed condition. Pearlite grains and ledeburite and secondary carbides are visible, 500 ×.

Now let us consider the effect of alloying elements on the positioning of transformation points and lines in the phase diagram for the iron-carbon alloys.

All alloying elements with the exception of aluminium and cobalt displace the transformation points S and E in the diagram to the left, this effect increasing with the carbon content in the steel. This may account, in particular, for the fact that high-speed steels containing only 0.70 to 0.95 per cent carbon have been found to belong to the carbide class, i. e., to lie to the right of point E in the phase diagram. As for the value of temperature A_3, it follows from the diagrams in Fig. 3 that the alloying elements causing the narrowing of the gamma-phase region (tungsten, molybdenum, vanadium, silicon, aluminium) raise the A_3 point, whereas those elements widening the gamma region (nickel, manganese) lower this point. As to the effect of chromium, it raises the A_3 point when contained in large amount, and depresses it when its proportion is small. The effect of alloying elements on the position of the A_1 point is similar.

From what has been said it is evident that the hardening temperature for chromium, tungsten and other steels alloyed with elements that narrow the gamma-phase region should be higher than that for carbon steels of the same carbon content, whereas that for nickel and manganese steels should be lower.

Chapter II

TRANSFORMATIONS IN STEEL ON HEATING

6. Transformation of Pearlite into Austenite

The transformation of the two-phase conglomerate of ferrite and cementite, known as pearlite, into the solid solution, or austenite, takes place at the A_1 temperature. For carbon steels the A_1 temperature on very slow heating is 723°C. Under real conditions of heating the pearlite transforms into austenite at higher temperatures ($A_1 > 723$°C).

Now let us consider how the transformation of pearlite into austenite will proceed in steel of eutectoid composition. For this purpose

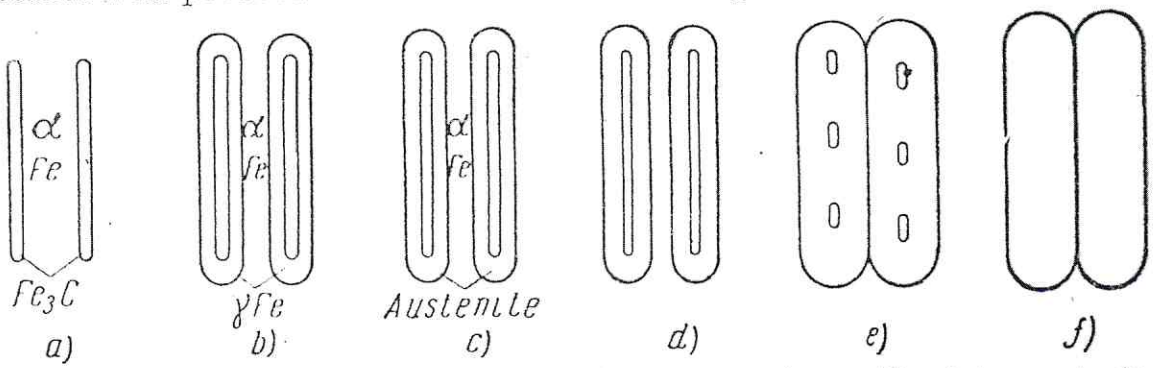

Fig. 9. Diagram illustrating the transformation of pearlite into austenite.

a part of a pearlite grain will be examined (Fig. 9, *a*). The cementite lamellas, or plates, are in direct contact with ferrite. The carbon atoms (contained in cementite) always tend to penetrate into the space lattice of the iron, this tendency becoming more intensive with an increase of temperature. In fact the solubility of carbon in ferrite, though insignificant in its absolute value, increases gradually with the rise of temperature (see Fig. 2).

On reaching the A_1 temperature the space lattice of alpha iron changes to that of gamma iron in the areas immediately adjoining the cementite lamellas (Fig. 9, *b*). Subsequently there takes place a process of gradual dissolution of the cementite in these areas of gamma iron just formed (Fig. 9, *c*). This results in the formation of the primary austenitic grain in which the surrounding ferrite dissolves, giving rise to a gradual growth of austenitic grains. At the same time a dissolving of the cementite lamellas in these grains takes place (Fig. 9, *d*). Both these processes continue until the complete

dissolving of ferrite and cementite is achieved. According to experimental data, the dissolving of the ferrite is completed before that of the cementite. In other words, at the moment when the dissolving of ferrite comes to an end, there are still small grains of pearlitic cementite remaining in the steel structure (Fig. 9, *e*). When these grains of cementite are dissolved, the steel structure becomes homogeneous in so far as its phase composition is concerned and is composed of austenitic grains only. But the grains of austenite remain chemically heterogeneous for some time, which is explained as follows: the concentration of carbon is high in those areas where cementite lamellas existed, while in peripheral regions it is low (Fig. 9, *f*). The carbon concentration is gradually levelled by diffusion in the whole austenitic grain. Chemically homogeneous grains of austenite are thus obtained.

From a consideration of this transformation pattern the conclusion may be drawn that an independent austenitic grain forms near to each cementite lamella. But as soon as these primary austenitic grains touch each other the process of their coalescence begins, through collective recrystallisation. This process proceeds simultaneously with the last stages of the transformation process. When the transformation is completed, there remain a number of austenitic grains instead of one grain of pearlite.

Such is the *mechanism* of the transformation of pearlite into austenite. We shall now discuss the *kinetics* of the transformation, i. e., the proceeding of the transformation process with time.

The kinetics of the transformation of pearlite into austenite is studied in the following manner. Small steel specimens are rapidly immersed into a bath heated to some constant temperature above that corresponding to equilibrium. Since the specimens are small, it can be accepted that they are very rapidly heated up to the bath temperature.

The aim of the experiment is to determine the speed of the transformation of pearlite into austenite. Readings are taken, beginning from the moment the specimen is put into the furnace. The degree of transformation can be estimated in several ways. For instance, a number of completely similar specimens may be immersed in a bath and then removed one by one at equal time intervals in order to be immediately quenched; their structure can then be examined and their hardness determined. Obviously the higher the degree of transformation, i. e., the larger the amount of pearlite transformed into austenite, the greater the quantity of martensite in quenched specimens and the higher their hardness. Thus, it is possible to observe the proceeding of the transformation process with time by the changes of structure and hardness or some other physical property.

The degree of the transformation of pearlite into austenite can be estimated in a still simpler manner, i. e., by the changes of the magnetic properties of specimens during the transformation process. Pearlite is magnetic, and austenite non-magnetic: as the transformation proceeds, the magnetic properties begin to decrease gradually until they reach zero on the completion of the process.

To record the changes of magnetic properties during transformation special devices such as magnetometers are used. A schematic diagram for one of them is shown in Fig. 10. The magnetometer consists of a powerful permanent magnet *1* which is drilled through axially, a tube *4* filled with molten salt being inserted into the hole. The tube is wound round with nichrome wire which is protected with heat-insulating matter *2*. So tube *4* represents a bath which can be heated up to a certain given temperature. A specimen to be tested, *8*, is fixed at one end of rod *3* which is made of non-magnetic metal. The phase transformations occur in the specimen *8*, immersed in the bath. The specimen is raised periodically to an interpolar space, position *5*. This changes the intensity of magnetic flux, recorded by a measuring ballistic coil *6* connected with the galvanometer *7*. The advantage of this method is that a single specimen is sufficient to illustrate the transformation process from beginning to end. Hardening is no longer necessary.

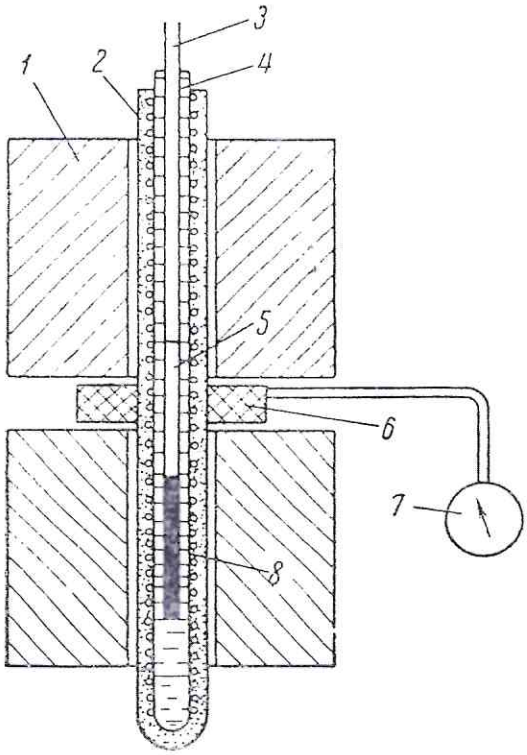

Fig. 10. Schematic diagram of a magnetometer.

Fig. 11 gives the results obtained in studying the kinetics of transformation of pearlite into austenite at several constant temperatures (i.e., the kinetics of the isothermal transformation). From this diagram it follows that the lower the temperature of isothermal heating, the longer is the transformation process.

From the data of this diagram another more convenient one can be drawn with time and temperature plotted as its coordinates (Fig. 12). The solid line in this diagram indicates the temperatures of the complete transformation of pearlite into austenite, if at constant temperature (isothermally). It is probable that the transformation process begins as soon as the steel has reached a temperature of isothermal heating. So if the steel is put inside a furnace with

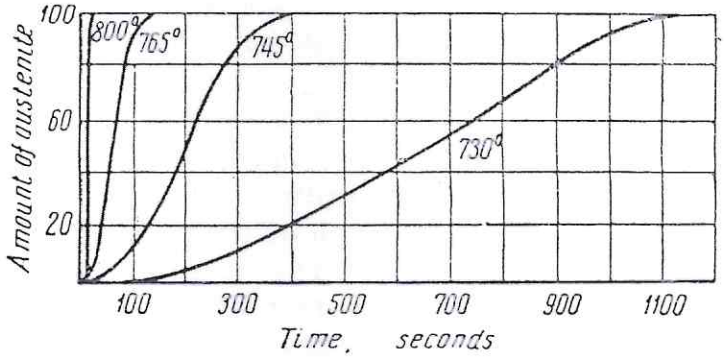

Fig. 11. Kinetics of the transformation of pearlite into austenite.

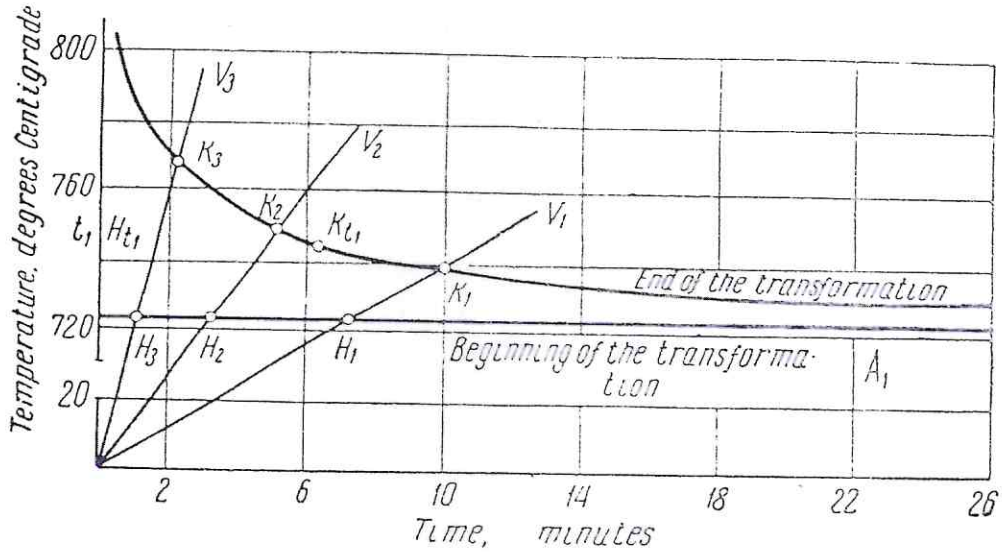

Fig. 12. Summary kinetic diagram of the transformation of pearlite into austenite.

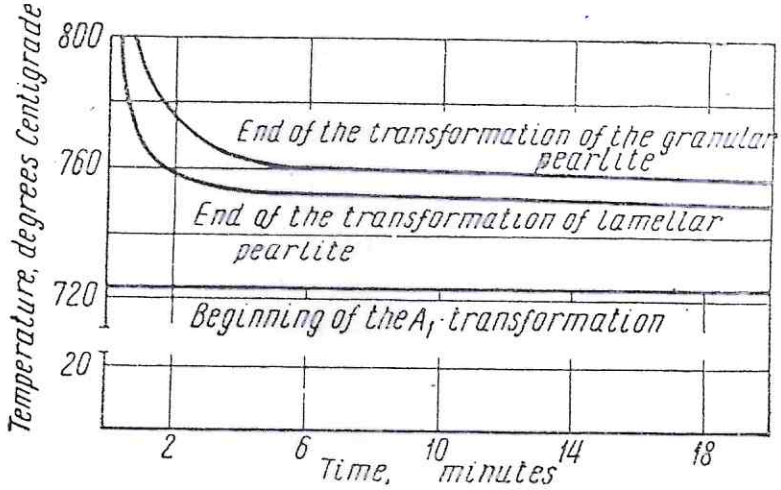

Fig. 13. Effect of the form of pearlite on the kinetics of its transformation into austenite.

temperature t_1 and has reached this temperature, the beginning of the transformation of pearlite into austenite is determined by the point H_{t_1} and the end by the point K_{t_1}. The duration of the transformation is determined by the section $H_{t_1}—K_{t_1}$.

From this diagram (Fig. 12) the kinetics of the transformation on continuous heating may be determined. There are three straight lines v_1, v_2 and v_3, which represent the heating operations at different heat rates; the steeper the line, the greater the heating rate. Thus, the lines v_1, v_2 and v_3 correspond to the low-rate, high-rate, and extremely high-rate heating, respectively. The transformation under continuous heating at any rate occurs at the same temperature, which is equal, or more precisely, somewhat higher than the A_1 temperature (the points H_1, H_2, H_3). The end of the transformation process is determined by the points K_1, K_2 and K_3.

It is therefore clear from this diagram that the transformation of pearlite into austenite on continuous heating occurs through a temperature range. The slower the heating, the narrower the temperature range for the transformation process, and on indefinitely slow cooling it is reduced to zero.

The diagram represents the isothermal transformation of lamellar pearlite into austenite. The granular pearlite transforms at a considerably lower rate (Fig. 13); according to the data of S. S. Steinberg the isothermal transformation of granular pearlite into austenite at 750°C takes five times longer than the transformation of lamellar pearlite at the same temperature.

7. Laws Governing the Growth of Austenitic Grains

At any temperatures from above the A_3 range up to the melting point, the structure of any steel is composed of austenitic grains. At temperatures above the A_3 range no phase transformations take place, but structural changes occur, i.e., the grains of austenite grow.

The interesting feature is not the growth of austenitic grains in itself but the fact that the grains of austenite in various steels possess different tendencies to growth. In steels of one kind the grains grow very rapidly (Fig. 14) and become very large after heating up to the temperature of 930°C. Such steels are called originally or inherently coarse-grained. In other steels the austenitic grains grow very slowly, and the steel retains the fine grain structure at the same temperature of 930°C. Such steels are said to have an original or inherent fine grain structure.

To estimate the tendency of austenitic grains to growth a temperature of 930°C is usually selected, being in practice the highest to which most steels are heated during heat-treatment. But this method of estimation, though it is quite adequate from the practical point of

view, is very conventional, as is evident from diagram (Fig. 14). On further heating above the temperature 930°C the austenitic grains of fine grain steel begin to grow rapidly, catching up and even overtaking in this respect the coarse grain steels. What is the cause of the different tendencies to growth of the austenite in various kinds of steel?

There is as yet no established opinion, but the most plausible explanation is as follows: the growth of austenitic grains is hampered by the finest ultramicroscopical particles of nitrides, oxides, and carbides distributed at the grain boundaries of austenite and forming as it were "barriers" between the grains. In fact it has been established that steels deoxidised by aluminium or modified by boron, titanium and other similar elements always possess inherent fine grain structure. But as the steel is heated these particles of nitrides, carbides, and oxides gradually dissolve in the austenite or coagulate (coalesce), the so-called "barriers" disappear and nothing remains to impede the growth of austenitic grains.

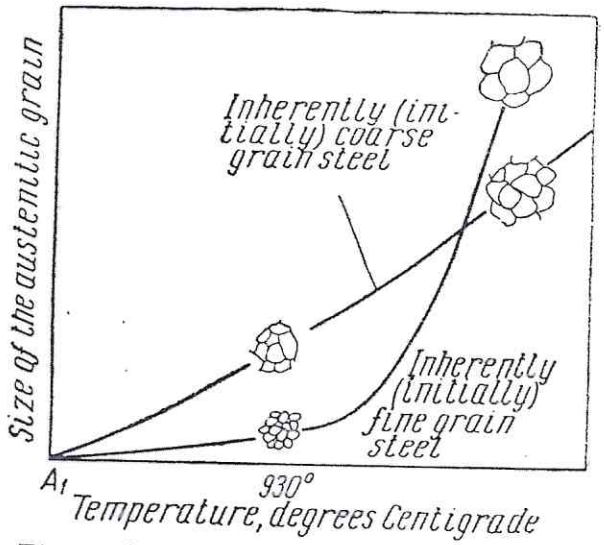

Fig. 14. Diagram illustrating the growth of austenitic grains.

8. Effect of Austenitic Grain Size on the Properties of Steel

The size of austenitic grains exerts a great immediate or indirect influence upon many properties of steel.

The austenitic grain size cannot produce any direct effect on the mechanical properties of steel, since the grains of austenite do not exist in steel structure at room temperatures, at which the mechanical properties are usually determined (this obviously does not apply to steels of the austenitic class). But an indirect influence of the austenitic grain size upon the mechanical properties of steel is unmistakable, i. e., the larger the austenitic grains, the larger under otherwise equal conditions will be the grains of ferrite and pearlite, the mechanical properties, particularly impact strength, being directly dependent upon the size of these grains.

The direct effect of the austenitic grain size on the structural changes in steel during its hardening proves to be much more substantial.

The transformation of austenite into pearlite begins mainly at the surface of the austenitic grains. Thus, the finer the grains of austenite, the greater will be their summary surface area and consequently the more rapidly will proceed the process of their change to pearlite. From this it follows that in hardening steels with a fine grain austenite structure higher cooling rates should be employed in order to suppress the tendency to transformation into pearlite.

From what has been said it is evident that the depth of hardening penetration, or the so-called hardenability, for two steels with the same chemical composition but with different sizes of austenitic grains will be different on cooling at the same rate; the larger the size of austenitic grains, the greater the hardenability of the steel.

When steel with a structure of coarse-grained austenite is hardened, coarser acicular martensite develops, and consequently the internal stresses are greater; hence the extent of deformation is larger and the forming of quenching cracks more probable. On the other hand, in steels with a structure of small austenitic grains the slip planes preceding the forming of acicular martensite are considerably less pronounced, and this accounts for the decrease in the internal stresses, the degree of deformation and the probability of crack formation.

9. Methods for Determination of the Austenitic Grain Size

In the State Standards there are four standard methods for determining the austenitic grain size in steels of the pearlitic class:

1) carburising—for hypoeutectoid steels;

2) forming of carbide network—for hypereutectoid steels;

3) high-temperature etching in molten salt bath—for steels of any carbon content;

4) high-temperature oxidation—for steels of any carbon content.

In the *carburising* method the specimen of hypoeutectoid steel is carburised at a temperature of 920°C to 940°C for a period of 8 hours, the time of heating up to this temperature not being counted, and then slowly cooled down to 600°C, the rate of cooling not exceeding 100 degrees per hour for carbon steels or 50 degrees per hour for alloy steels. The rate of further cooling is of no importance. The austenitic grains develop on the carbide network of the carburised part of the specimen.

In the method of determining the austenitic grain size by means of *carbide network* the specimen of hypereutectoid steel is heated up to the same temperature and is kept at this temperature for 3 hours, being cooled afterwards to 600°C at a rate of 80 to 100 degrees per hour.

In the method of *high-temperature etching* in molten salt bath a microsection of a steel specimen is heated in a furnace with protective gas atmosphere, or in a common furnace, being protected from the oxidation by the iron chips, borax, charcoal powder and waste carburiser, or in bath of molten sodium carbonate. The heating is accomplished at a temperature of 920°C to 940°C for a period of 3 hours, the hot-etching test being then conducted in a molten mixture consisting of equal quantities of barium chloride ($BaCl_2$), sodium chloride (NaCl) and calcium chloride ($CaCl_2$). The etching test is accomplished at a temperature of 930°C for 2 to 5 minutes, the cooling in kerosene being then carried out until all the salt is washed off. The austenitic grains usually become visible without the additional etching of a microsection.

In the *oxidation* method a microsection is heated in a furnace with an oxidising atmosphere at the same temperature for 3 hours and then is quenched in water. The austenitic grains become visible on the network of oxides after additional grinding, polishing and etching of microsection by a 15 per cent solution of hydrochloric acid for a period of 2 to 10 minutes.

The *ferrite network* may also be used as a means of determining the austenitic grain size in hypoeutectoid steels containing 0.3 to 0.5 per cent carbon.

The size of austenitic grain made visible by any one of the above-mentioned methods is determined through microscopic examination at a magnification of 100 diameters by comparing the structure of a microsection with the scales shown in the standard referred to. Altogether 12 grain numbers have been established:—1, 0, 1, 2, etc., up to 10 inclusive.

Vacuum metallography offers very interesting possibilities for the determination of austenitic grain size. Heating cross-sections of the steel specimens at high temperatures in a vacuum (10^{-4} to 10^{-5} mm of mercury column) produces a vapourisation of atoms from the surface of the cross-section which is the most intensive at the grain boundaries (selective evaporation), the boundaries of the austenitic grains being thus made visible. The austenitic grains distinguished by this method can be viewed by means of an ordinary microscope after cooling the cross-section.

10. Changes in Steel Structure on Heating Not Caused by Phase Transformations

When unhardened steel of the pearlitic class is heated at a temperature below the A_1 range, an increase of the solubility of carbon in ferrite is noticed. But as this increase is very insignificant (0.01 per cent at room temperature and 0.02 per cent at 723°C),

it cannot produce any marked changes in steel structure. However, on heating up to the A_1 temperature considerable structural changes in steel may occur. These changes can be of two kinds:

1) if the steel has been hardened by cold working, the recrystallisation process takes place on heating at a temperature below the A_1 range;

2) if the steel is kept for a long time at high temperatures (but below the A_1 point), the processes of coagulation and spheroidising of cementite occur.

The recrystallisation process consists in that at certain temperatures above the so-called recrystallisation point the refined and elongated grains in the steel structure first gradually assume a regular geometrical (equiaxial) form (primary recrystallisation) and then grow larger as a result of the growth of some grains at the expense of the others (collective recrystallisation). The steel structure was made up of both pearlite and ferrite grains before the heating, as well as after it. Thus, its phase composition does not change during recrystallisation, whereas its structure does undergo the substantial changes. The properties of the steel change correspondingly, in that the tensile strength, yield point and hardness decrease, while those of elongation, reduction of area (characterising the ductility of the steel), and impact strength rise.

The spheroidising process consists in that the cementite plates tend to assume a globular form, and the coagulation process consists in the gradual disappearance of the small cementite grains and in the simultaneous coarsening of the larger cementite grains. The processes of spheroidising and coagulation (coalescence), proceeding simultaneously, are explained as follows. The small cementite grains are surrounded by the ferrite, which can dissolve carbon (though in small quantities). The solubility of carbon in the ferrite is not uniform: the carbon concentration is always higher in areas immediately adjoining small cementite grains than in those adjacent to larger ones. The solubility of different sides of cementite plates in ferrite is also not uniform. Therefore the carbon concentration in ferrite is unequal, and the presence of the concentration differential causes the diffusion of carbon, proceeding from those areas where its content is higher (i. e., near the small grains of cementite and narrow sides of the cementite plates) to those areas where its concentration is lower (i. e., near the larger cementite grains and broad sides of the cementite plates). The depletion in carbon of the first areas is made up by the gradual dissolution of small grains and the narrow sides of the cementite plates in ferrite, whereas the increase in carbon percentage of the ferrite areas near the larger cementite grains or broad sides of the cementite plates gives rise to a separation of excess carbon in the form of the cementite

molecules from the ferrite and to their precipitation on larger grains and the broad sides of the cementite plates. As a result of these processes the small grains and lamellas of cementite gradually disappear, while the larger grains progressively become coarser.

The processes of recrystallisation, spheroidising and coagulation are based on the common physical law known as the second principle of thermodynamics, i. e., each system (particularly each phase) spontaneously tends to diminish its store of free energy. This tendency is expressed in the reduction of the phase surface area. This is precisely the nature of the processes of recrystallisation, spheroidising, and coagulation, i. e., with primary recrystallisation and spheroidising the grains acquire an equiaxial (geometrically regular) form similar to the globular one, the sphere of ball possessing the smallest surface area possible of all bodies of the same volume; during the collective recrystallisation and coagulation (coalescence), the grains are coarsening, the surface area of one grain being always less than the sum-total of the surface areas of several grains having the same summary volume and similar shape.

Chapter III

TRANSFORMATIONS IN STEEL ON COOLING

11. Mechanism of the Transformation of Austenite into the Ferrito-Cementite Conglomerate

Austenite is a solid solution of the carbon, or of carbon and other elements in gamma iron. If steel is held at temperatures above the A_3 point for a sufficiently long period of time, the atoms of carbon are uniformly distributed in the space lattice of the gamma iron. But this uniformity is an average or statistical one. The carbon atoms may continuously move inside the space lattice, leaving some crystalline nuclei and penetrating into others. The average statistical uniformity of distribution of the carbon atoms in austenite is therefore continuously broken; in some areas of the austenite the proportion of the carbon atoms is greater than average, while in others it is less. Such breaches of uniformity of concentration are called *compositional fluctuations*. It is true that the regions with greater or lesser quantity of carbon atoms exist for only a very small time interval. The same continuous heat movement of carbon atoms that causes the fluctuation of the chemical composition tends to equalise the concentration of carbon atoms. Those portions of austenitic grains that formerly possessed a high concentration of carbon atoms subsequently become "normal" in this respect, whereas other portions of austenitic grains which were "normal", become momentarily supersaturated with carbon atoms; the third kind of grain areas, on the other hand, are depleted of them. This process goes on continuously and constantly. Among the portions of austenitic grains in which the concentration of carbon atoms becomes higher than average as a result of this fluctuation, there are also areas in which the carbon concentration is 6.7 per cent, i. e., up to the percentage of carbon in the cementite.

Thus, in the austenitic grains areas enriched in carbon up to the cementite concentration are constantly being formed on the one hand, and depleted of it so as to form a pure iron, on the other hand. On cooling the steel below the A_1 point, i. e., on supercooling it, some areas become centres or nuclei for the crystallisation of the ferrito-cementite conglomerate. What is the mechanism of this transformation?

Let us suppose that in some section of the austenitic grain there has been formed an area in which the carbon concentration is 6.7

per cent. In this area cementite molecules will form, i. e., the crystalline nucleus of cementite will arise. The adjoining areas of the austenitic grain will thus become depleted of carbon, with pure iron resulting. As the temperature is below 910°, in these areas there will occur a change of the space lattice of gamma iron into that of alpha iron, i. e., crystalline centres or nuclei of ferrite will be formed. The growth of each of the nuclei (cementite and ferrite) will stimulate the further growth of another nucleus; thus, the growth of the small crystal of cementite causes the further depletion of the adjacent areas of austenite in carbon atoms, thereby favouring the growth of the small crystals of ferrite. The growing ferrite crystals will reject the carbon atoms to their boundaries, thus providing the material for growth of the cementite crystals. This process, begun in some one place, will progressively occur in still greater volumes of the austenitic grain, resulting in the formation of a grain of the ferrito-cementite conglomerate. The process of formation of the ferrito-cementite conglomerate grains usually begins simultaneously in several areas of an austenitic grain.

Such are the basic features of the mechanism of transformation of austenite into the ferrito-cementite conglomerate. One more question remains to be clarified. Fluctuations in the austenite composition take place at any temperature: why do these fluctuations not give rise to the formation of the ferrito-cementite aggregate at all temperatures, but only at temperatures below 723°? This is explained by the fact that the free energy of austenite at temperatures above 723° (the A_1 point) is less than that of the ferrito-cementite conglomerate, and consequently austenite is stable, whereas the ferrito-cementite aggregate is unstable. At the A_1 temperature the free energies of austenite and the ferrito-cementite conglomerate are equal, being thus in equilibrium; for this reason the transformation process of austenite into the ferrito-cementite aggregate cannot take place. Yet at temperatures below the A_1 the free energy of ferrito-cementite conglomerate is less than that of the austenite, the former being therefore stabler, thus accounting for the unhindered change of austenite to the ferrito-cementite aggregate.

12. Kinetics of the Transformation of Austenite into Ferrito-Cementite Conglomerate

The study of the kinetics of the austenite transformation into ferrito-cementite mixture, i. e., the determination of the temperature effect upon the rate of this transformation has been experimentally found to be most easily carried out not on continuous cooling but at a constant temperature, i.e., under *isothermal* conditions.

Experimentally this is accomplished in the following manner. Small specimens of the steel to be tested are heated to some temperature above the A_3 point, and after some period of holding at that temperature in order to be heated throughout they are rapidly transferred into a bath which has been heated previously to some temperature below the A_1 point. Consequently, the process of transformation of austenite into ferrito-cementite conglomerate will take place when the specimen is in this bath. As has been found for steel, the best method of recording the transformation of austenite into

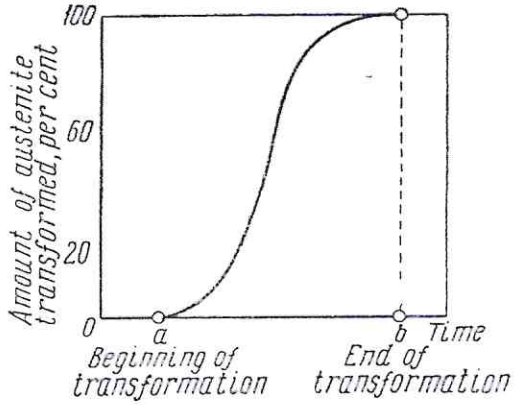

Fig. 15. Kinetic curve showing the isothermal transformation of supercooled austenite at a given temperature.

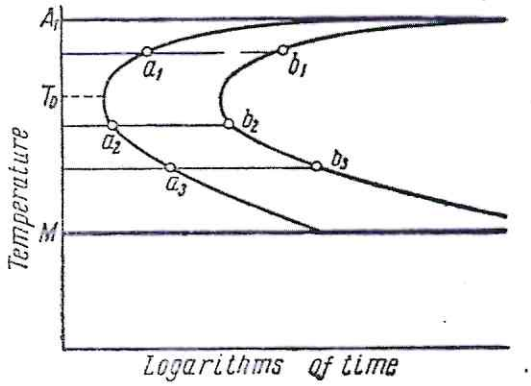

Fig. 16. Summary kinetic diagram showing the isothermal transformation of supercooled austenite.

ferrito-cementite conglomerate is the magnetometric one. The *kinetic curves* representing the change of austenite to ferrito-cementite aggregate have been constructed by means of this magnetometric method.

Fig. 15 gives the curve showing the *isothermal transformation* of austenite into ferrito-cementite conglomerate in eutectoid steel at a given temperature. From this curve it may be seen that the transformation process does not begin as soon as the steel has attained the given temperature, i.e., not immediately after the specimen has been transferred into the furnace of the magnetometer. The transformation does not occur for some time, which is measured by the length Oa. The transformation begins at point a and reaches completion at point b after a lapse of time determined by the length ab. The steel has austenitic structure up to point a, and the transformation occurs between points a and b, i.e., austenite and ferrito-cementite conglomerate exist simultaneously in the steel structure; after point b has been passed only the ferrito-cementite aggregate remains.

Let us construct such isothermal curves for different temperatures and then proceed as follows. Let us take temperature and time

as coordinates (Fig. 16), draw a number of isotherms (horizontal lines) corresponding to temperatures, for which these isothermal curves have been constructed (Fig. 15), mark on these isotherms points *a* (beginning of the transformation) and *b* (end of the transformation), then connect all points *a* and *b* with even lines. A summary *graph illustrating the isothermal transformation* of austenite will then be obtained. In order to limit the length of the diagram on the abscissa, time is not plotted on it, but a logarithm of time.

The left curve on this diagram determines the *beginning* of the transformation of austenite into ferrito-cementite conglomerate,

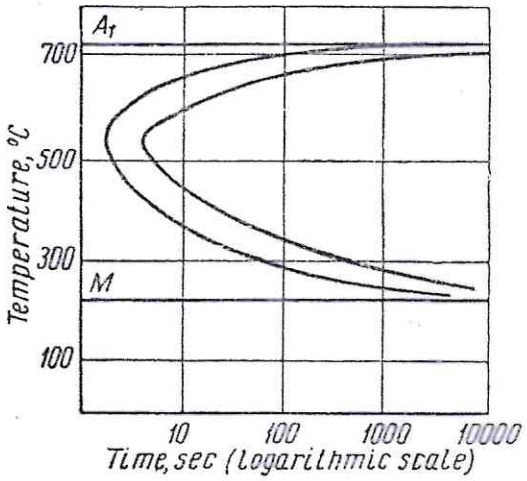

Fig. 17. Diagram showing the isothermal transformation of supercooled austenite in steel of grade У8.

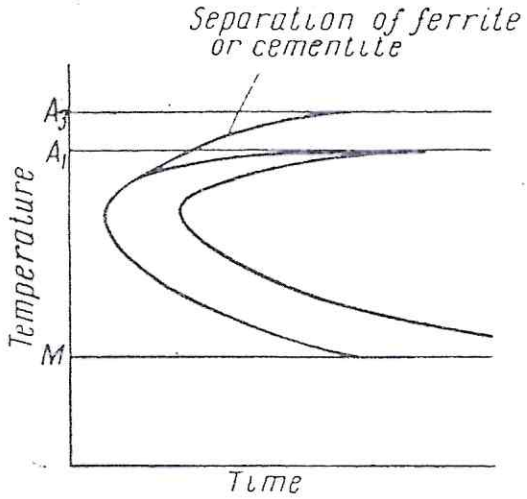

Fig. 18. Diagram showing the isothermal transformation of supercooled austenite in hypoeutectoid or hypereutectoid steels.

while the right one shows the *end* of this change. The austenite exists in the region to the left of the left curve, the ferrito-cementite mixture in the field to the right of the right curve. The transformation process proceeds in the area located between these two curves. This means that in this area both austenite and the products of its decomposition, i. e., ferrito-cementite conglomerate exist simultaneously.

The line *M* represents the temperature of martensitic transformation. This line and the martensitic transformation itself will be discussed more amply in the following paragraphs.

A number of extremely important conclusions may be drawn from the diagram of the isothermal transformation of austenite (Fig. 16).

At temperatures below the bottom critical point A_1, which corresponds to equilibrium, the supercooled austenite may exist for a length of time determined by the left transformation curve. The *degree of stability of the austenite* varies at different temperatures of supercooling. The austenite is least stable when supercooled at

temperatures within less than 100 to 150° of the equilibrium temperature A_1. This minimum stability of the austenite is determined by the point T_0 of the inflexion of the left curve.

The austenite is found to be more stable at temperatures of supercooling higher than the T_0 point, as well as lower than this range. The stability of austenite is highest at temperatures but slightly below the A_1 point, on the one hand, and at those approaching the M temperature on the other hand.

The real diagram of the isothermal transformation of supercooled austenite for the steel of grade У8 is represented by Fig. 17. This diagram is similar to the schematic one which has just been considered. The diagram for the hypoeutectoid and hypereutectoid steels (Fig. 18) is similar, too, but with the addition of only one curve which represents the separation of an excess phase, i. e., of ferrite in hypoeutectoid steels or of cementite in hypereutectoid steels.

13. Structure and Properties of Ferrito-Cementite Conglomerate

The isothermal transformation of supercooled austenite at any temperature in the A_1-M range gives rise to the formation of an aggregate composed of ferrite and cementite. But depending upon the transformation temperature, the structure and properties of this ferrito-cementite conglomerate are substantially different.

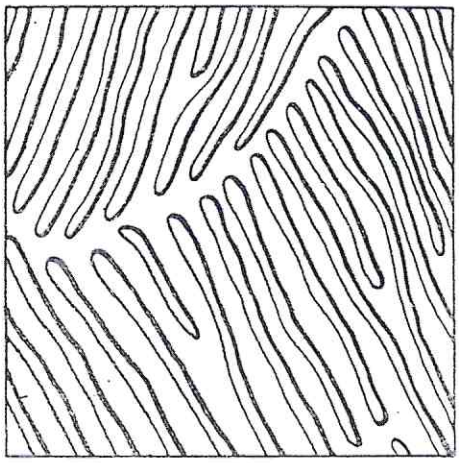

Fig. 19. Structure of lamellar pearlite at magnification of 2,500. Sorbite has the same structure at 5,000 ×, and troostite at 10,000 ×.

The ferrito-cementite aggregates formed at different temperatures of supercooling differ primarily in the degree of dispersion (refinement) of both phases which are constituents of this conglomerate, i. e., of ferrite and cementite. The lower the transformation temperature, the more dispersed are both phases. At high temperatures of transformation (at about 700°) the ferrito-cementite conglomerate becomes sufficiently distinct (differentiated) (Fig. 19). Such a well-defined aggregate composed of ferrite and cementite is called pearlite. At somewhat lower temperatures of transformation (at about 650°) the ferrito-cementite conglomerate is more dispersed. In this case it is called *sorbite*. At still lower temperatures of transformation (at 600° and down to the M range) the ferrito-cementite con-

glomerate becomes so dispersed that its structure can hardly be distinguished at any magnifications by an optical microscope. Only the photomicrographs made with an electronic microscope reveal that its structure is similar to that of pearlite and sorbite. Such highly dispersed ferrito-cementite aggregate is known as *troostite*.

The ferrito-cementite aggregates differ from each other not only in degree of dispersion of constituent phases but also in the form of the cementite phase. The cementite phase in ferrito-cementite conglomerate may have either lamellar or globular (grained) form. The formation of the cementite phase of any kind depends upon the temperature to which the austenite has been heated prior to its transformation. It has been found that when austenite is heated up to comparatively high temperatures (exceeding the A_1 point by 100° or more) the ferrito-cementite conglomerate has a lamellar structure irrespective of the transformation temperature (Fig. 20, *a*). If the austenite has been heated to comparatively low temperatures (exceeding the A_1 point by not more than 50 to 60°) and the degree of supercooling has been relatively small, the cementite phase will have a grained structure (Fig. 20, *b*). Even at greater degrees of supercooling in this case the structure will become lamellar.

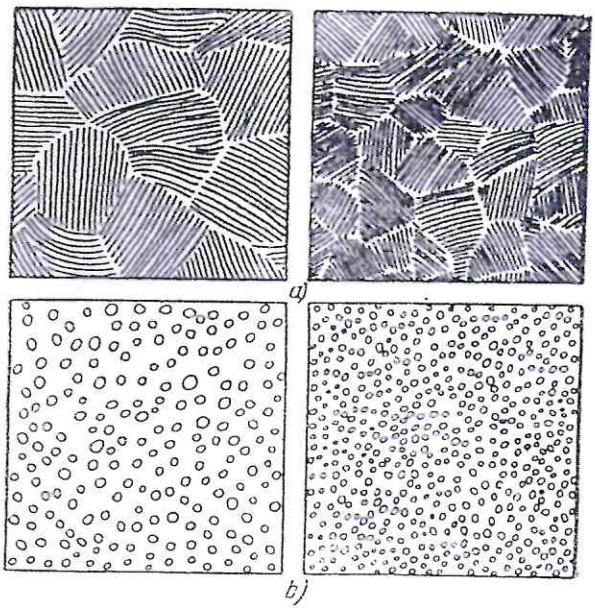

Fig. 20. Structure of the ferrito-cementite conglomerate in carbon steel in relation to the temperature of heating of the initial austenite and the transformation temperature.

a — lamellar structure: heating up to 900° and transformation at temperatures 710° (left) and 670° (right); *b* — grain structure: heating up to 780° and transformation at temperatures 710° (left) and 670° (right), 500×.

Ferrito-cementite aggregate of sufficiently distinctive (differentiated) structure with a cementite phase of grained form is called *grain pearlite* or, as some scientists prefer to call it, *grain cementite*. Ferrito-cementite conglomerates of more dispersed structure with cementite in grain form have not been given a special name, and are called sorbite or troostite like the dispersed ferrito-cementite aggregates of lamellar structure.

The forming of a cementite phase of any kind in the ferrito-cementite conglomerate is explained by the degree of uniformity of the orig-

inal (primary) austenite. The more uniform austenite (which corresponds to higher temperatures of heating it) is decomposed into an aggregate of lamellar structure. At lower temperatures of heating the uniformity of structure is not attained; apart from the austenite, undissolved particles of cementite remain, which determine the grain form of the separated austenite.

The transformation of the supercooled austenite in the temperature range from A_1 to about 500° (Fig. 16) gives rise to the formation of polyhedrons or spherolites of ferrito-cementite

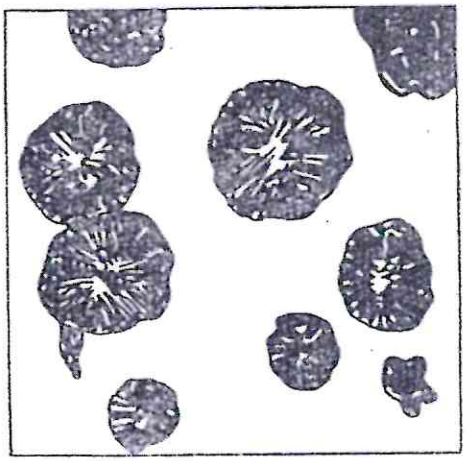

Fig. 21. Globules of troostite formed in the upper zone of the supercooled austenite decomposition, 500 ×.

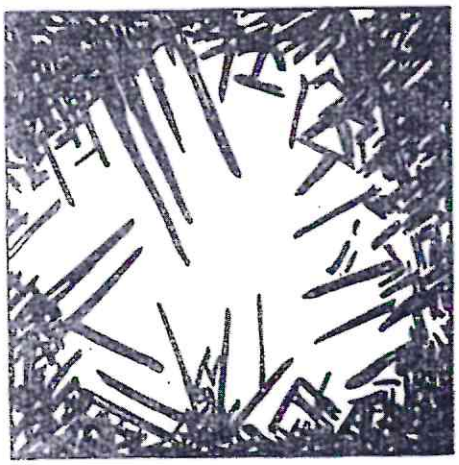

Fig. 22. Structure of acicular troostite, 500 ×.

conglomerate (Fig. 21), while the transformation in the temperature range from about 500° to M leads to the formation of a distinctive acicular structure (Fig. 22) similar to that of martensite. By studying the structure of these needles, in particular by means of an electronic microscope, it has been found that it is composed of the same aggregate of ferrite and cementite as is the troostite. Therefore this acicular structure is often called the *acicular troostite* (or sometimes *bainite*).

Besides the difference in microgeometry, there are also other differences between troostite and acicular troostite. First, the carbon content in the ferrite phase is somewhat different; the carbon content in the ferrite of troostite corresponds to that determined by the phase diagram, while the carbon content in the ferrite of acicular troostite is in excess of that for a condition of equilibrium. In other words, when the acicular troostite is formed not all the carbon contained in the austenite separates from the space lattice of iron. So the process of formation of the acicular troostite is both of diffusion and non-diffusion nature.

The second difference between troostite and acicular troostite is that the austenite is transformed into troostite through the generation of grains and their subsequent growth, whereas the change of austenite to acicular troostite is accompanied by hardly any growth of needles.

These two peculiar features of the formation of acicular troostite make it more approximate to martensite. According to these differences in the formation mechanism and in the microgeometry of the products of austenite decomposition into ferrite and cementite, the whole temperature range from A_1 to M (designated as Ar) is divided into two sections: 1) the upper range (or upper step), from A_1 to about 500°, where pearlite, sorbite and troostite are formed; and 2) the lower range (or lower step), at about 500° to M, where the acicular troostite is formed. Transformation in the upper step is designated as Ar' and in the lower step Ar''. The properties of ferrito-cementite aggregates forming at different temperatures are different. The more dispersed the ferrito-cementite conglomerate, the higher are its strength, elasticity and hardness. As to plasticity and ductility, they improve first as the degree of dispersion increases, and begin to deteriorate after the maxima have been reached. Sorbite has the best values for plasticity and ductility (under otherwise equal conditions), these being somewhat higher than those of pearlite and considerably higher than those of troostite.

The Rockwell "C" hardness of eutectoid steel may be related to the degree of dispersion of the ferrito-cementite conglomerate by the following figures:

pearlite	10-20
sorbite	20-30
troostite	30-45
acicular troostite	45-55

14. The Martensitic Transformation

When steel is cooled to the M temperature (see Fig. 16) or to still lower temperatures, the austenite changes not into the ferrito-cementite conglomerate but into martensite.

The transformation of austenite into ferrito-cementite aggregate consists, as has already been amply explained, of the following two simultaneous processes: 1) change of the space lattice of gamma iron to the space lattice of alpha iron; 2) separation of the majority of the carbon atoms from the space lattice of gamma iron and their precipitation into an isolated cementite phase. The second process consisting in migration of the carbon atoms inside the space lattice is thus a diffusion process. The first process, the change of the space lattice, is a non-diffusion process.

With high degrees of supercooling to comparatively low temperatures (200 to 300°) the space lattice changes freely. The second process, i. e., the separation of most of the carbon atoms from the space lattice of iron and their precipitation in the form of the cementite molecules, does not take place at low temperatures of supercooling for very obvious reasons that at comparatively low temperatures such as these the rate of carbon diffusion is very slow. The carbon atoms remain, therefore, in the space lattice of iron.

What is obtained as a final result? The fact that the carbon atoms remain in the space lattice of iron means that the martensitic structure obtained is a solid solution. But since martensite has the space lattice of alpha iron, it may be deduced that martensite is a solid solution of carbon in alpha iron.

The nature of martensite was completely unexplained not long ago. From the easily established fact of the martensitic structure of hardened steel being magnetic, the conclusion may justly be drawn that the space lattice of martensite is similar to that of alpha iron. On the other hand, it was experimentally proved long ago that the electrical resistivity of martensite is considerably higher than that of the ferrito-cementite conglomerate of any degree of dispersion. High values of electrical resistivity are characteristic of solid solutions. It therefore seems apparent that martensite is a solid solution of carbon in alpha iron. But this conclusion does not conform with the well known fact that alpha iron is hardly capable of dissolving the carbon. So contradictions arise. Many theories have been advanced to explain the nature of martensite. In books on metallography published in the twenties and even thirties of the current century a description of these theories may be found.

The Soviet scientists N. Y. Selyakov, G. V. Kurdyumov and N. T. Gudtsov were the first to discover the nature of martensite in 1927; they examined the structure of martensite by means of X-rays. A. P. Gulyaev rightly claims that the works of these scientists may be considered as important as those of P. P. Anosov and were epoch-making in the field of steel metallography.

In their work, corroborated by many later investigations, it has been established that: 1) martensite is a single-phase system; 2) it has a tetragonal space lattice; and 3) the parameters of the space lattice are a linear function of the carbon content in it (Fig. 23). From this it has been deduced that the martensite does not possess a stoichiometrical composition, i. e., it is not to be considered as a chemical compound. Yet what else can a single-phase system of varying chemical composition be, if not a solid solution?

The mechanism of martensitic transformation may be represented as follows. On rapid cooling, high internal stresses exceeding the

low yield point of the austenite develop in the austenitic grains. This results in plastic deformation of the austenitic grains, i. e., some planes of the space lattice move in relation to each other, bringing about a slip. The space lattice becomes distorted and partly destroyed, and to replace it the iron atoms form another lattice which is more stable at temperatures below the A_3 point, i. e., the space lattice of alpha iron.

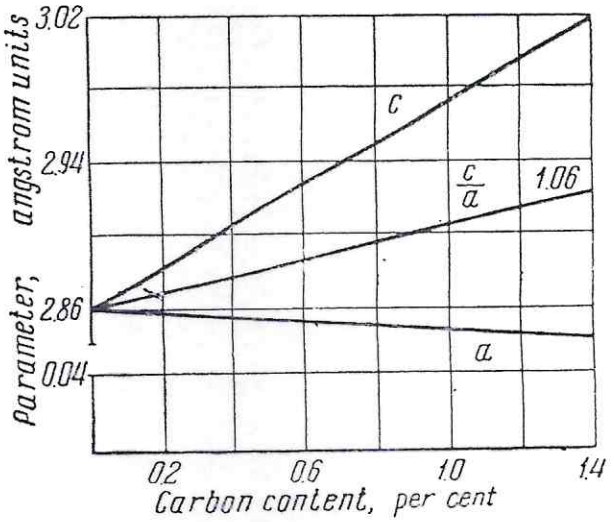

Fig. 23. Relation of parameters of the martensite space lattice to the carbon content.

Thus, the elongated and narrow area of the austenitic grain, located along the line of slip with the space lattice of alpha iron, is actually a plate of martensite, which appears in the microsection as a long and thin needle. The acicular form is a very characteristic microstructural feature of martensite (Fig. 24).

The first needles of austenite pierce the entire austenitic grain almost throughout (Fig. 25, *a*). The rate of formation of martensitic needles is enormously high. The high velocity in taking photomicrographs (800 shots per second) of martensitic transformation, made by A. P. Gulyaev, did not show any growth of martensitic needles; a needle appears at once in some of the stills, though there is no sign of it on the preceding one, but the time interval between the two shots is only about 0.001 second. It may therefore be accepted that the martensitic needles appear practically instantly. The martensitic needles formed do not grow in length or breadth but remain invariable. As the martensitic transformation proceeds further new needles form (Fig. 25, *b*, etc.), all subsequent needles becoming shorter and shorter.

From the above discussed mechanism of martensitic transformation it follows that this process is non-diffusion, the chemical com-

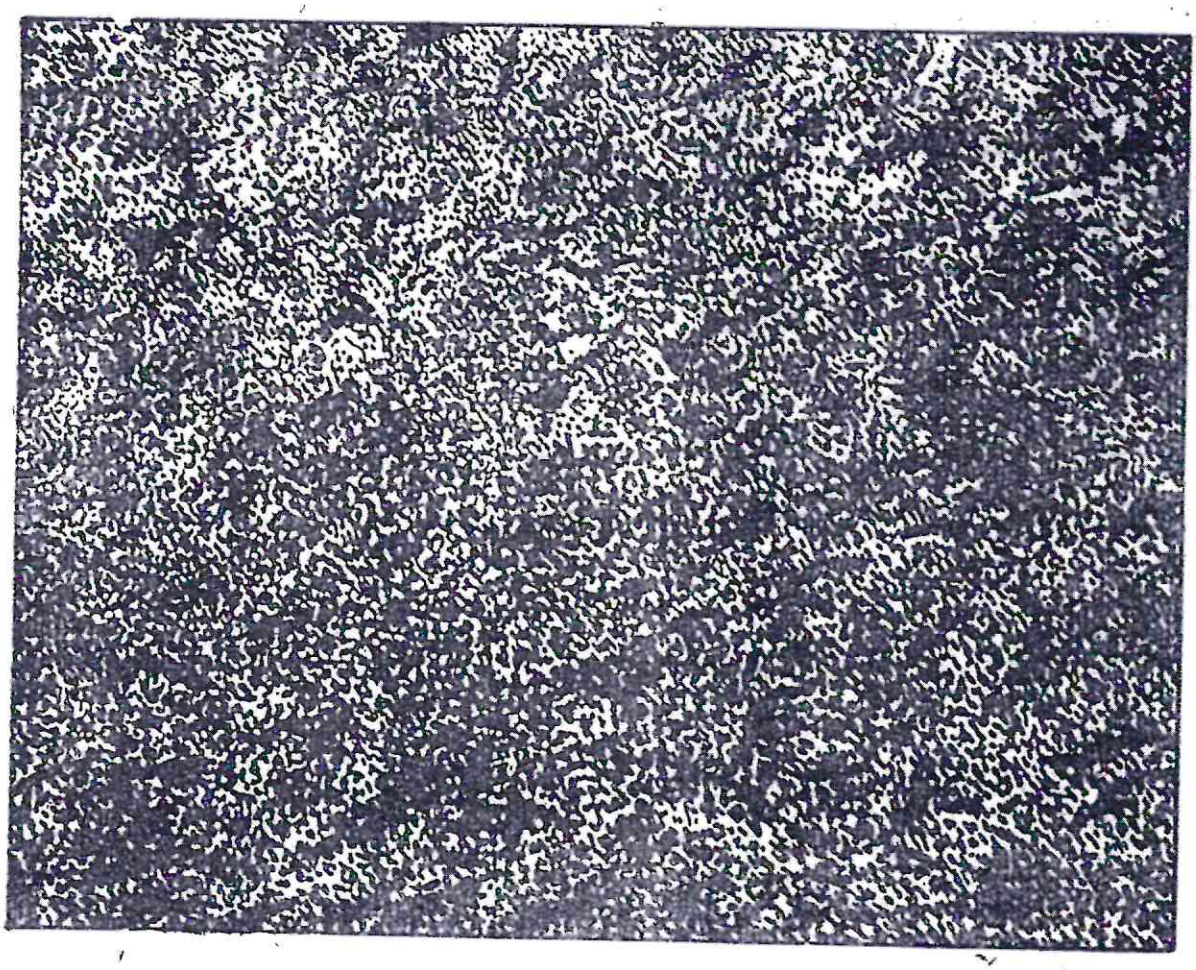

Fig. 24. Structure of martensite in hardened steel of grade У8, 500 ×.

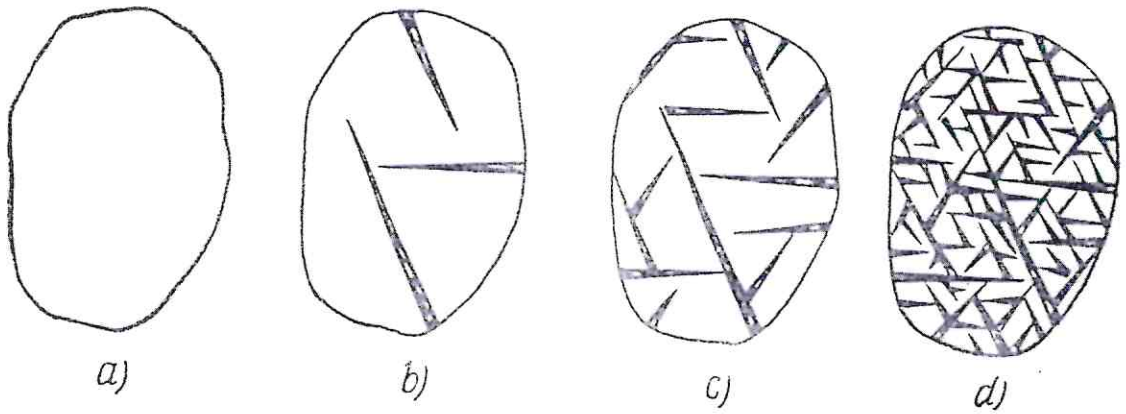

Fig. 25. Scheme showing the formation of martensitic structure. The figure represents one austenitic grain at several progressively lowering temperatures (from left to right) below the *M* point. After the entire volume of the grain has been filled with martensitic needles, the areas of residual austenite (light coloured) remain between them.

position of austenite and that of martensite forming from it being completely identical.

Unlike the ferrito-cementite transformation, the martensitic transformation can hardly proceed isothermally. When steel is held at a constant temperature below the M point, the martensitic transformation ceases very rapidly. The resulting structure is made up of austenitic grains pierced by some quantity of martensitic needles. The temperature of the steel should be lowered progressively so as to enable the process to develop further. As the temperature is lowered, new stresses are set up in the austenitic grains, leading to the formation of new martensitic needles.

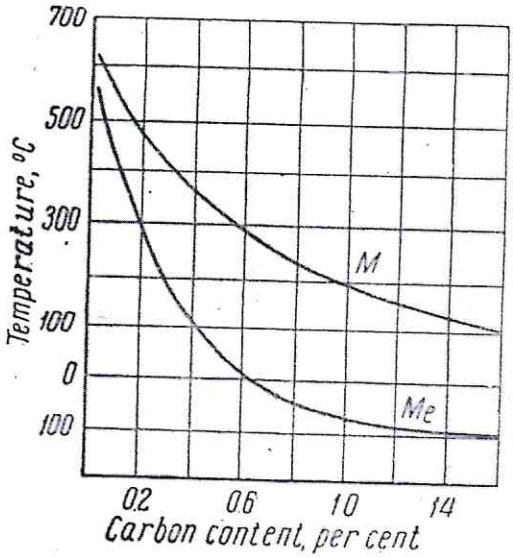

Fig. 26. Diagram showing relation of temperatures of beginning and completion of martensitic transformation in carbon steel to the carbon content.

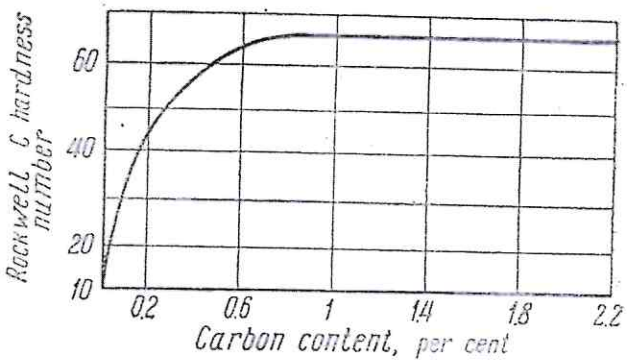

Fig. 27. Diagram showing relation of the hardness of iron-carbon alloys quenched so as to obtain the martensitic structure to the carbon content in these alloys (diagram as given by I. S. Gayev according to data of various investigators).

Thus, the martensitic transformation is accomplished through a temperature range. The temperature ranges M-M_e (corresponding to the beginning and end of the martensitic transformation respectively) are determined by the carbon content in the steel (Fig. 26), i. e., the higher the proportion of carbon in the steel, the lower the temperatures of the beginning and end of the martensitic transformation.

From the diagram in Fig. 26 it is clear that the martensitic transformation in steels with a carbon content exceeding 0.5 per cent is completed at temperatures below room temperature. This fact enables one to draw two very important conclusions. First, the structure of any hardened steel containing carbon in excess of 0.5 per cent is made up, if it is not cooled below room temperature, not only of martensite but also of some amount of austenite which

has not been transformed. The higher the carbon content in the steel, the greater the amount of austenite will be. This austenite is called *residual*. Therefore when the structure of a steel containing carbon in excess of 0.5 per cent is said to be composed of martensite, the "residual austenite" is also implied. And secondly, in order to increase the amount of martensite in the structure of hardened steels with carbon in excess of 0.5 per cent and to increase thereby their hardness, it is sufficient to cool them below room temperature (to sub-zero temperatures). This is the main point of treating steel by cold, which will be dealt with in greater detail in Chapter VIII.

Some of the residual austenite remains in the structure of hardened steel even when this is cooled below the M_e temperature at which the martensitic transformation ends. This is because when austenitic grains are cooled to still lower temperatures they are broken by the martensitic plates into separate "sections", inside which the austenite is compressed from all sides. The specific volume of martensite being greater than that of austenite, this fact accounts for the hampering of the martensitic transformation by compression of the austenite in these "sections" on all sides.

Martensite may be obtained in the form of either coarse or fine needles. The degree of the acicularity of the martensite depends upon the size of the original austenitic grains: the larger the austenitic grains, the larger will be the needles of martensite. Since the austenitic grain size depends upon the excess of heating temperature above the A_3 point, the size of the martensitic needles may be used as a basis for determining the degree of steel overheating when it is heated for quenching. In the structure of normally hardened hypereutectoid tool steel the martensitic needles cannot usually be detected because they are extremely fine.

The characteristic properties of martensite are its very high hardness (Fig. 27) and extremely low impact strength. The explanation of both these properties of martensite is the same, i. e., the considerable distortions of the space lattice of martensite, which hinder plastic deformation. The higher the carbon content in martensite, the greater the distortions of the space lattice, and, hence, the greater will be the resulting hardness. This conclusion is corroborated by Fig. 27.

15. Transformation During Continuous Cooling

The study of the kinetics of the transformation of austenite under isothermal conditions has considerably reduced the experimental difficulties, for it made possible the elimination of the influence of the cooling rate. Because of this many laws governing this trans-

formation have been established, the most important of them being already considered.

However, under real conditions of heat-treatment the transformations occur in most cases *not isothermally* but on *continuous cooling*. Therefore the task is to employ the diagram of isothermal

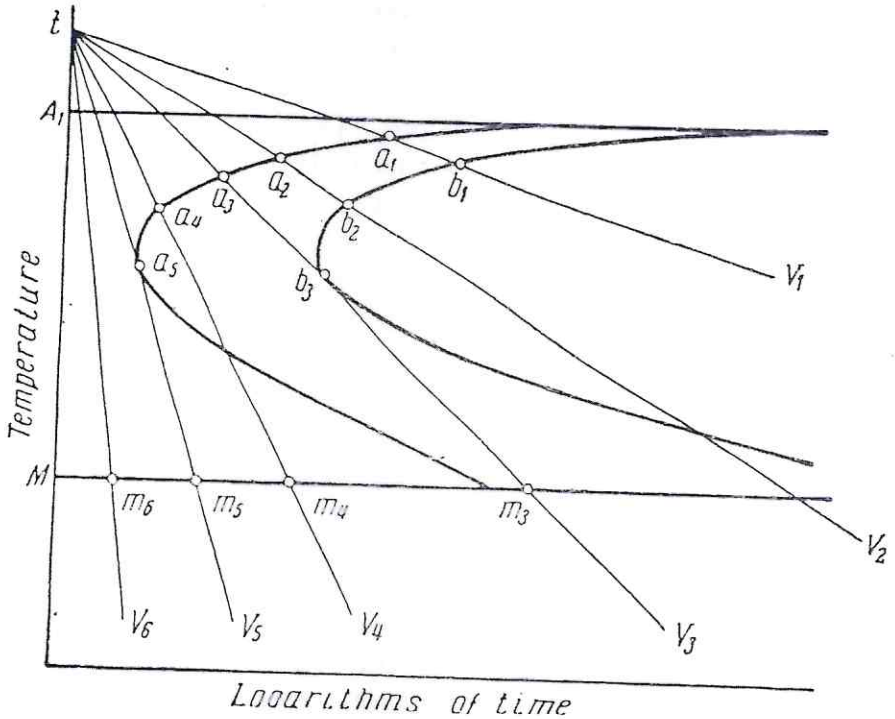

Fig. 28. Plotting of curves, illustrating continuous cooling, on the diagram of isothermal transformation of supercooled austenite.

transformation of supercooled austenite for the study of the transformations that take place under continuous cooling. This problem has been resolved by plotting the curves for continuous cooling on the diagram of the isothermal transformation of supercooled austenite. The lines for continuous cooling plotted on the diagram with coordinates temperature-time in logarithmic scale are represented by curves, but in order to simplify the plotting these curves will be considered as straight lines.

Several specimens of eutectoid steel are heated to a temperature above the A_1 point. This temperature is marked in the diagram of isothermal transformation of supercooled austenite as point t (Fig. 28). Then the steel specimens are cooled at various cooling rates. The cooling processes are represented by the inclined straight lines, which are the more steep, the faster the cooling rate. The slowest cooling rate (for annealing) is represented by the straight line v_1, the slightly more rapid cooling rate (for normalising) by the straight line v_2, the still more rapid cooling rate (for oil quenching)

by the straight lines v_3 and v_4, and the most rapid cooling rate (for water quenching) by the straight lines v_5 and v_6.

On very slow cooling the straight line v_1 crosses both transformation curves as follows: the curve of beginning of transformation is intersected at point a_1, and the curve of completion of transformation at point b_1. This means that on slow cooling the austenite completely transforms into a ferrito-cementite aggregate. Since the transformation occurs at the highest temperatures (for the temperature range A_1-M), the ferrito-cementite conglomerate becomes the least dispersed and sufficiently distinctive. This is pearlite.

On more rapid cooling (as in normalising) the straight line v_2 also intersects both transformation curves. This means that even in this case the austenite completely decomposes into ferrito-cementite conglomerate; but as the transformation range a_2-b_2 corresponds to lower temperatures, the resulting aggregate is more dispersed. This is sorbite.

At any cooling rate not exceeding v_3 the curves illustrating the cooling process intersect both transformation curves, which implies the complete decomposition of the austenite into ferrito-cementite aggregate.

At a cooling rate faster than v_3, e. g., at a cooling rate v_4, the straight line v_4, illustrating cooling, intersects only the curve corresponding to the beginning of transformation (the a_4 point), and does not cross the curve representing the end of transformation. This means that the ferrito-cementite transformation begins but does not reach completion. In other words part of the volume of the austenitic grains undergoes decomposition into ferrite and cementite, while for another part of this volume there is insufficient time for it to be transformed into a ferrito-cementite aggregate.This remaining part of the austenite, which has not been decomposed, undergoes the transformation into martensite on reaching the M temperature (the m_4 point). Thus, the structure of steel cooled at the rate v_4 consists partly of troostite and partly of martensite. Such a structure is peculiar to all steels cooled at a rate faster than v_3, but slower than v_5. For a carbon steel this cooling rate is equivalent to quenching in oil.

At any cooling rate faster than v_5 the transformation of austenite into ferrite and cementite does not take place at all, and the austenite completely transforms into martensite (the points m_5, m_6). For carbon steel this cooling rate corresponds to the quenching in water.

In analysing the transformations taking place during continuous cooling one common failure, which is usually committed, must be avoided, when the diagram is considered in a formal way. This concerns the straight lines v_3 or others which are more sloping (be-

tween v_3 and v_1). These straight lines also intersect the M line showing the martensitic transformation; e. g., the v_3 line crosses the M line in the m_3 point. But from this it cannot be deduced that at such cooling rates the martensitic tranformation does take place in the steel. In fact at the cooling rates equal to or less than v_3 the straight lines representing the cooling intersect both curves, and hence in

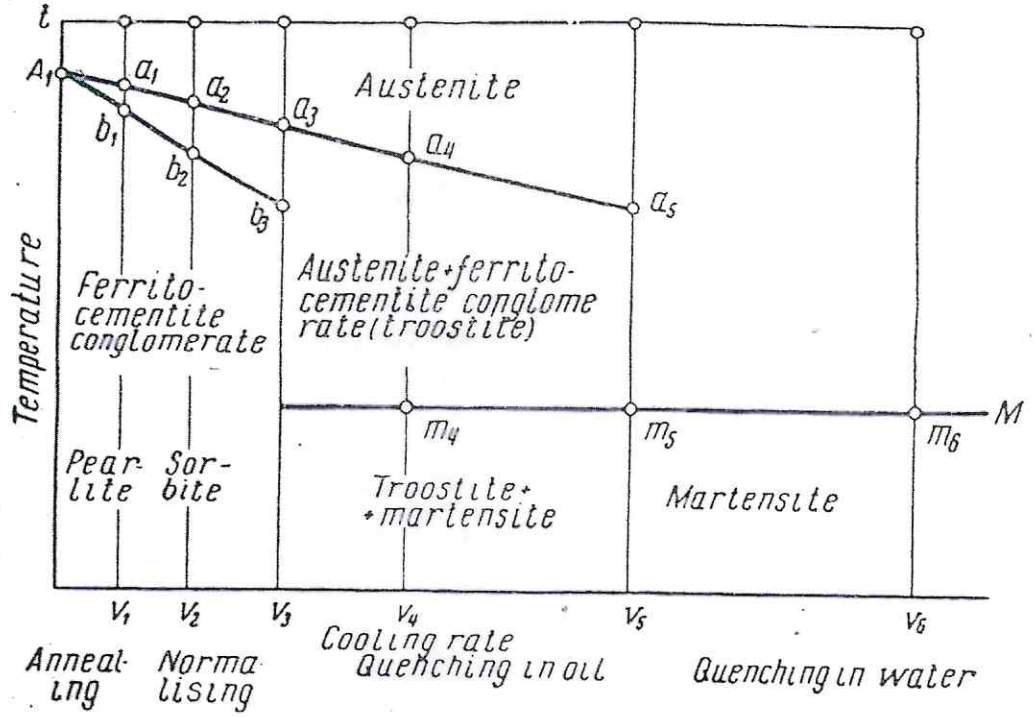

Fig. 29. Diagram showing relation of austenite transformation temperatures to the cooling rate (the letter symbols are the same as in Fig. 28).

steels thus cooled the ferrito-cementite transformation reaches completion. No austenite not already transformed remains, and consequently there is *nothing* to be transformed to martensite at the M temperature (the m_3 point). So under conditions of continuous cooling the m_3 point, and others located in the straight line M to the right of the m_3 point, have no physical sense.

The cooling rate v_3, the least at which the martensitic areas appear in the steel structure, is called the *low critical quenching rate*. It is evident that the minimum critical quenching rate corresponds to the straight line illustrating the cooling, which is a tangent of the curve indicating the end of transformation. The cooling rate v_5, being the least at which the ferrito-cementite conglomerate does not remain in the steel structure, which is composed only of martensite (and unavoidable residual austenite), is called the *upper critical quenching rate*. Again it is clear that the upper critical quench-

ing rate corresponds to the straight line for cooling, which is a tangent of the left curve indicating the beginning of the pearlitic transformation.

It should be noted here that by the critical cooling rate is usually understood precisely this upper critical quenching rate.

The relation between the structures obtained on continuous cooling and the cooling rates can easily be made much clearer if the temperatures of beginning and completion of the martensitic transformation are plotted on the diagram with the temperature and cooling rate as coordinates (Fig. 29). From this information the diagram is easily understood.

16. Peculiarities of the Transformations of Austenite in Alloy Steels

In alloy steels the mechanism of the transformation of austenite into martensite and ferrito-cementite conglomerate does not qualitatively differ at all from that of the same transformations in carbon steels.

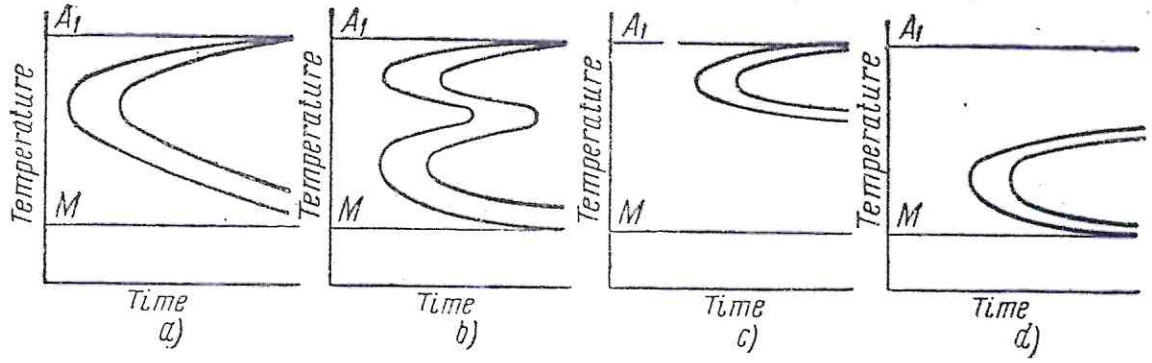

Fig. 30. Main types of diagrams showing the isothermal transformation of the supercooled austenite in alloy steels.
a — with only one step of transformation; *b* — with two steps of transformation; *c* — with only one upper step of transformation; *d* — with only one lower step of transformation.

The mechanism of the first (two-phase) transformation differs only in that this transformation results in the formation not of an aggregate of ferrite and cementite but of a conglomerate made up of ferrite and carbides, because many alloying elements form complex carbides with cementite.

As to the kinetics of transformation, there are differences in this respect between carbon and alloy steels. For steels alloyed with elements that are not carbide forming (nickel, copper, aluminium) and containing small proportions of manganese (up to 2 to 3 per cent) the diagram for isothermal transformation of the supercooled austenite is of the same pattern (Fig. 30, *a*) as that for carbon steels,

i. e., in the diagram there is only one minimum representing stable austenite.

For steels alloyed with carbide-forming elements (chromium, tungsten, molybdenum, vanadium) the diagram for isothermal transformation of the supercooled austenite possesses two minima indicative of the stable austenite (Fig. 30, *b*). The first (upper) minimum corresponds to the Ar'-transformation, i. e., the transformation of austenite into ferrito-carbide aggregate. The lower minimum corresponds to the Ar''-transformation, i. e., the transformation of austenite into acicular troostite.

All alloying elements, both the carbide forming and those which do not form carbides (except cobalt) move the transformation curves to the right, this effect being the more pronounced the higher the content of these elements in the steel (Fig. 31). In other words, the alloying elements increase the stability of austenite.

Most alloying elements (chromium, tungsten, molybdenum, vanadium, silicon, aluminium) raise the temperature corresponding to the minimum stability of austenite, while some of them (nickel, copper, manganese) depress it (Fig. 31).

And finally, all alloying elements (with the exception of aluminium and cobalt) depress the temperature of the martensitic transformation, this effect being the more pronounced the higher the content of these elements in the steel. With a considerable content of alloying elements in the steel the temperature of the martensitic transformation may be below zero. This means that in such steels stable austenitic structure is obtained at room temperature.

The diagrams for the isothermal transformation of austenite in some complexly alloyed steels are yet more complex and unusual (Fig. 32). In particular in the diagram of the isothermal transformation of austenite for steels of type 18ХНВА there is no upper minimum indicative of the austenite stability; for this reason the very long periods of holding are necessary for the Ar'-transformation to occur. This means, of course, that in such cases only the Ar''-transformation (the formation of the acicular troostite) and the martensitic transformation take place at ordinary cooling rates.

These differences in the kinetics of transformation of austenite in alloy steels cause a number of peculiar features in the heat-treatment of such steels.

Let us compare the diagrams of the isothermal transformation of austenite for three steels of pearlitic class (Fig. 33): plain carbon steel (Fig. 33, *a*), low-alloy steel (Fig. 33, *b*) and high-alloy steel (Fig. 33, *c*). These diagrams differ from each other in that the curves of the ferrito-carbide transformation are moved to the right to a greater degree for the highly alloyed steels than for low-alloyed

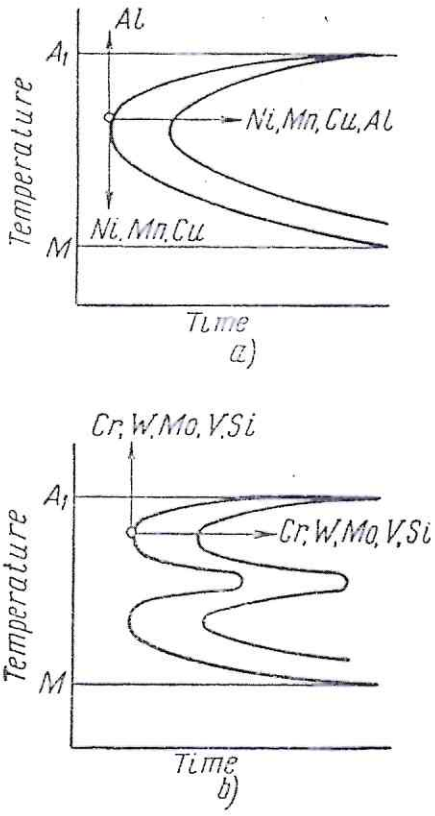

Fig. 31. Effect of alloying elements upon the diagram of isothermal transformation of supercooled austenite.

a — in steels with only one step of transformation; *b* — in steels with two steps of transformation.

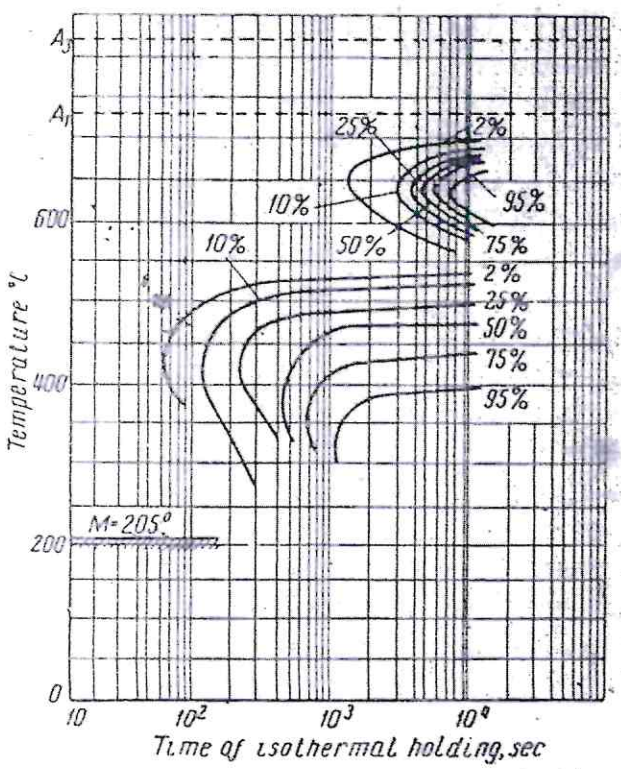

Fig. 32. Diagram showing the isothermal transformation of austenite in steel of grade 5XHB (A. A. Popov, V. S. Sagaradze, S. E. Khorzov and E. F. Vostrikova).

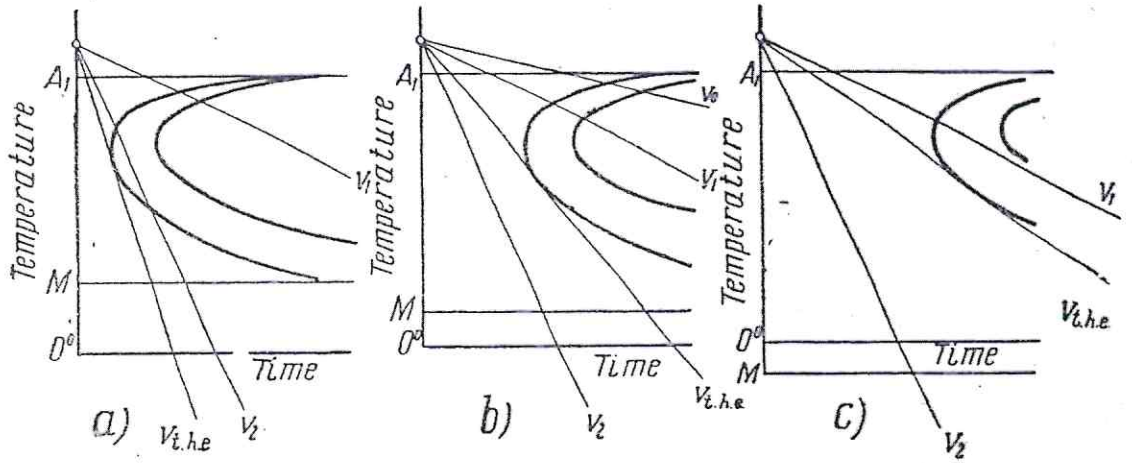

Fig. 33. Comparison between the diagrams of isothermal transformation of austenite for three differently alloyed steels.

ones (supposing, for the sake of simplification, that in these three steels there is no Ar''-transformation).

Let us draw the straight lines showing the cooling in annealing v_1 on all these three diagrams. As is sufficiently well known, in the case of carbon steels the straight line for annealing intersects both transformation curves at the very top. This means that a clearly distinctive pearlite with a moderate hardness will be obtained in annealing the carbon steel. Since in many cases the specific purpose of annealing is to reduce the hardness of steel, this aim will be thus obtained.

The annealing of low-alloy steels presents quite another picture. The straight line for annealing v_1 (Fig. 33, *b*) intersects both transformation curves, but does so at lower temperatures. This means that as a result of the usual annealing of low-alloy steels a ferrito-carbide aggregate of a more dispersed structure (sorbite or even troostite) will be obtained, and therefore the hardness of annealed low-alloy steel will be markedly higher than that of annealed carbon steel; in most cases this is absolutely inadmissible and the annealing must therefore be accomplished at very low cooling rates (5 to 15 degrees per hour). It is impossible to accomplish such retarded annealing in a switched-off furnace which is allowed to cool down freely, for any heat-treating furnace cools at considerably higher rates. Therefore the retarded annealing is usually done in a furnace which, though it is cooling, has at the same time been warmed up to a certain extent. The straight line indicative of this annealing process is shown in Fig. 33, *b*, being marked by the symbol v_0.

In annealing highly alloyed steels the straight line characteristic of the usual annealing process at best intersects only one left curve, which indicates the beginning of the transformation (Fig. 33, *c*). This means that in the structure of annealed highly alloyed steels martensite will be found.

To cause pearlitic transformation in highly alloyed steels such low cooling rates that cannot be obtained in practice are necessary. How, in such cases, can the reduction of hardness of a highly alloyed steel be achieved? It may be achieved not by annealing but by the high tempering, in which the spheroidising and coalescence of carbides occur, i. e., the reduction of the degree of dispersion.

The degree of stability of austenite increases with the concentration of alloying elements; this fact, being the cause of many difficulties experienced in annealing, proves to be very favourable for hardening.

On the same three diagrams there are three straight lines v_2 corresponding to quenching in oil. The straight line v_2 for carbon steel (Fig. 33, *a*) intersects only the left curve indicating the beginning

of transformation. This means that both troostite and martensite will be found in the structure of a carbon steel quenched in oil. The same straight line v_2 on the diagram for the low-alloy steel (Fig. 33, *b*) passes to the left of the left transformation curve. This means that an alloyed steel quenched in oil has a martensitic structure. Yet quenching in oil has a tremendous technological advantage over water quenching, since the internal stresses and strains usually set up by the rapid cooling are considerably reduced thereby; consequently, the danger of warping and formation of quenching or water cracks is materially reduced.

Highly alloyed steels quenched in oil (some of them even after normalising) possess an austenitic structure. This is explained by the fact that the M point of martensitic transformation lies below room temperature (Fig. 33, *c*).

17. The Real Speed of Cooling and the Critical Cooling Rate of Quenching

Cooling is the most important part of any heat-treating operation. The cooling rate determines the kind of structure resulting, and thus the specific properties of the steel obtained. The cooling rate is especially important in hardening. In order to obtain a martensitic structure on hardening the real speed of cooling must be higher than the critical quenching rate.

If the numerous but secondary factors are neglected the *real cooling speed* mainly depends upon the *weight of article* to be hardened, its *surface area* and the *nature* and *temperature* of the quenching or *cooling media*. Therefore it is wrong to say that the faster speed of cooling is always obtained on quenching in water, while the lower on quenching in oil. The speed of cooling of a massive article in water may be lower than the critical cooling rate, and martensitic structure will not be obtained. On the other hand, the speed of cooling of a small article in oil may be higher than the critical cooling rate in quenching, not only on the surface but also in the central portion, so that the article will be hardened throughout and martensitic structure will result (Fig. 34).

When quenching in water is said to be necessary for obtaining martensitic structure this refers only to relatively small articles and tools, the weight of which does not exceed a few kilogrammes. It is usually precisely such articles and tools that have to be quenched, and for them the real speed of cooling in water is always higher than the critical cooling rate of quenching.

The *critical cooling rate of quenching* also depends upon a number of factors. The chemical composition of steel is the main factor determining the value of the critical cooling rate of quenching. The

relation of the critical cooling rate of quenching carbon steels to the carbon content is shown in Fig. 35.

For some alloy steels the values of critical cooling rate (according to Esser) are shown below:

Carbon content, per cent	Alloying element, and its proportion	Upper critical cooling rate in quenching, degrees per second
0.42	1.12 Ni	450
0.40	4.8 Ni	85
1.09	0.53 Cr	300
1.02	1.04 Cr	100
0.92	2.00 Cr	25
1.19	3.03 Cr	20

The critical cooling rate of quenching depends to a considerable degree upon the *hardening temperature*; the higher the heating temperature for hardening, the more uniform becomes the austenite,

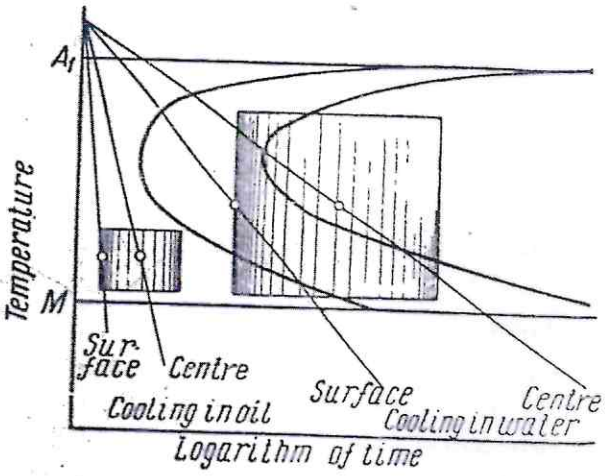

Fig. 34. Effect of a weight (mass) of articles made from the same steel upon the type of transformation of austenite; the smaller article is cooled in oil, the larger one in water.

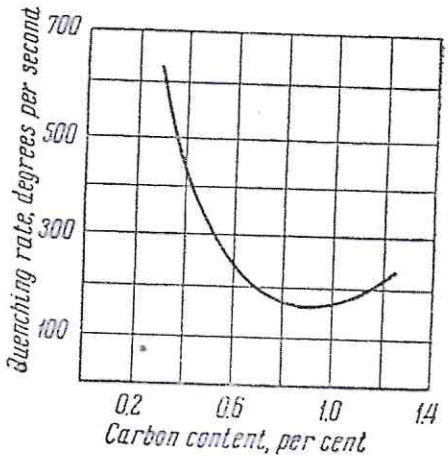

Fig. 35. Diagram showing the relation of the critical cooling rate of quenching carbon steels to the carbon content in them (after S. S. Steinberg).

the larger its grains, the stabler it will be and, consequently, the lower its critical cooling rate of quenching. The critical cooling rate of quenching depends to a certain extent upon the *metallurgical nature of the steel*; the cleaner the steel with regard to various non-metallic inclusions, the less, consequently, the number of centres or nuclei, in the steel structure facilitating the *Ar*-transformation the stabler the supercooled austenite in the upper step, and thus the slower the critical cooling rate of quenching.

The critical cooling rate in quenching does not depend, however, upon the weight of articles. Unlike the real cooling rate, *the critical speed of quenching is an intrinsic property of the steel itself.*

18. Hardenability

When articles are quenched the cooling rate decreases from the surface to the interior, being the highest at the surface and the lowest in the central portion.

If the cooling rate in the central portion of an article is in excess of the critical one for the given grade of steel, the article will be

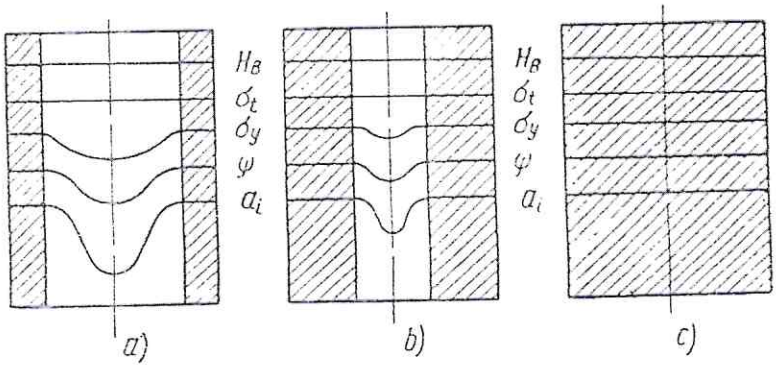

Fig. 36. Diagram showing the change of mechanical properties through the cross-sections of articles made of three steels of different hardenability after they have been quenched and tempered so as to obtain equal hardness (according to A. P. Gulyaev).

hardened through its entire cross-section, i. e., hardened throughout. But if the cooling rate in the centre of the section is lower than the critical one, the central portion remains unhardened.

The hardenability is of great importance for obtaining high and, what is more important, uniform properties across the whole cross-section of an article.

Suppose that three equal-size articles made of three different grades of steel possessing different hardenability have been hardened (Fig. 36), and that two of them (*a* and *b*) have not been hardened throughout (the hardened zone is shaded), whereas the third (*c*) has. The hardened zones will have the martensitic structure, while the structure of non-hardened zones will be troostite or even sorbite.

Let us carry out high tempering of all three articles at the same temperature. The martensite of the hardened zones will then change to the so-called tempering sorbite. As to the non-hardened zones, in high tempering the quenching sorbite hardly changes its structure, but in the quenching troostite the spheroidising processes

will partially occur, though in general its structure will be nearer to that of quenching sorbite than to that of tempering sorbite.

Therefore after high tempering the hardened zones will be composed of the tempering sorbite and the unhardened zones of the quenching sorbite. There is a very substantial difference between these two kinds of sorbite. Quenching sorbite is of lamellar structure (see Fig. 20, *a*), whereas tempering sorbite is of grained structure (see Fig. 20, *b*). This distinction in structure of the two sorbites causes a substantial difference in their properties. There is almost no difference between quenching sorbite and tempering sorbite (tempered so as to obtain the same hardness) with regard to their values of tensile strength and elongation, but in yield limit, reduction of area, and especially impact strength, tempering sorbite is markedly superior. Therefore the central unhardened portions of cross-sections for articles *a* and *b* made of steels with poor hardenability possess lower ductility and toughness than article *c* made from steel of high hardenability, though in all three articles the hardness will be exactly the same.

Thus, all parts which must possess high and uniform mechanical properties across the whole cross-section must be manufactured of steels with high hardenability.

According to the degree of hardenability, all structural steels may be divided into four groups:

1) steels of poor hardenability; hardening throughout is obtained in cylindrical specimens with diameters not exceeding 15 mm; this group includes the carbon steels;

2) steels of medium hardenability; these harden throughout with diameters up to 35 mm; this group includes the low-alloy steels such as 30Г2, 15Х, 30Х, 33ХС, 40ХС, 40ХФ and others;

3) steels of increased hardenability (diameter up to 75 mm); this group includes more comlexly alloyed steels, such as grades: 40ХН, 30ХСГ, 30ХМА and others;

4) steels of high hardenability (diameter up to 100 mm); this group includes the complexly alloyed steels containing three or more alloying elements, for example steels of grades 20ХН3А, 12Х2Н3МА, 18ХНВА and others.

Hardenability, being a function of the critical cooling rate of quenching, depends upon the same factors as the critical quenching speed. Thus, the effect of the *content of alloying elements* in steel is clear. Hardenability, like the critical cooling rate of quenching, depends upon the *austenitic grain size*. Originally coarse-grain steels harden to a greater depth than originally fine-grain steels. The hardenability of steel may be somewhat improved by raising the *quenching temperature*; this is explained as follows: the

coarsening of the austenitic grains and the dissolution of non-metallic centres or nuclei (incipient phases) increase the stability of the supercooled austenite, reduce the critical cooling rate of quenching and, consequently, improve the hardenability.

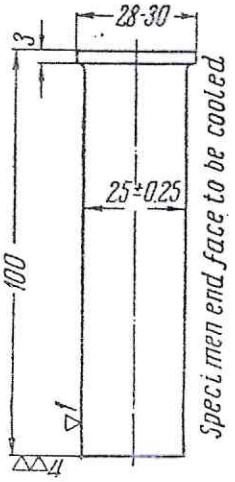

Fig. 37. Specimen to be used for determining the hardenability of steel with the standard end-quench testing (according to State Standard).

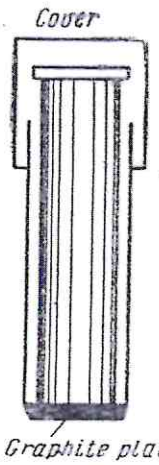

Fig. 38. Method of heating specimen without oxidation for the end-quench test.

The hardenability of steel is determined by various methods. The simplest one, the so-called end-quenching method, is specified by the U.S.S.R. State Standard. To determine the hardenability of a steel, a special cylindrical specimen is made (Fig. 37). This specimen is heated in a furnace in such a manner as to prevent the oxidation and decarburisation of its end-face, as is shown in Fig. 38. After being maintained at a proper temperature for a period of 30 min, the heated specimen is rapidly transferred (the time for this operation should not exceed 5 seconds) to the quenching unit (Fig. 39), where it is immediately quenched by water spraying.

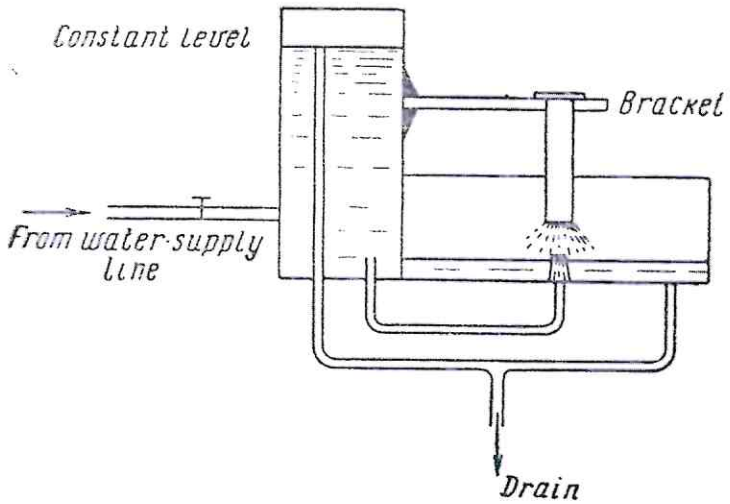

Fig. 39. Sketch of the unit for determining hardenability of steel with the end-quench test.

Two opposite sides of the cylindrical surface of the specimen are then ground so as to obtain two longitudinal areas sunk

by 0.2 to 0.5 mm, and at these surfaces the Rockwell "C" hardness is determined, the measurements beginning at a distance of

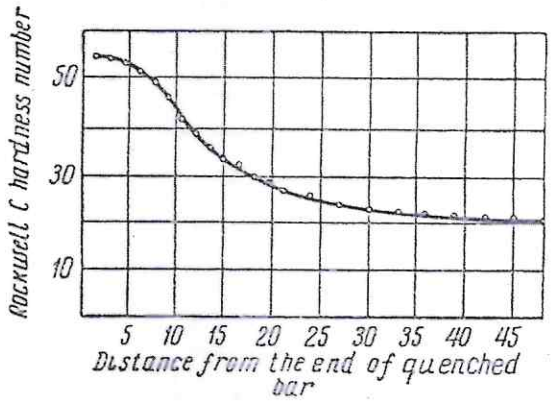

Fig. 40. The hardenability curve.

48 mm from the hardened end. The first ten measurements in the direction of the hardened end are made every 3 mm, and the

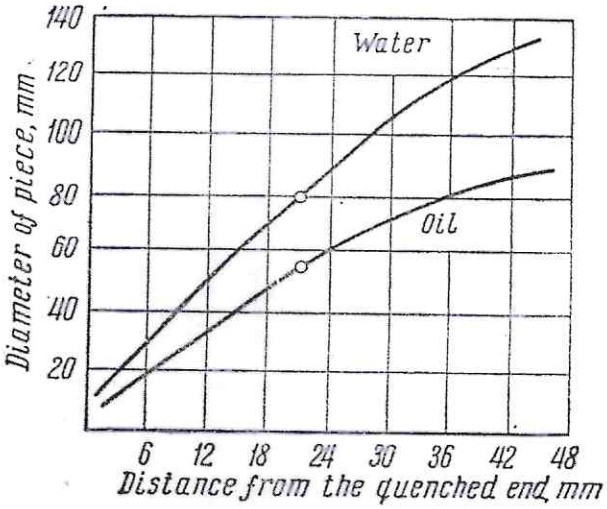

Fig. 41. Diagram illustrating the graphical method for determining, on the basis of the hardenability curve, the diameter of a cylindrical article having a given hardness after quenching.

Example: From the data obtained in the end-quench test it has been established that the hardness $40R_C$ is located at a distance of 21 mm from the quenched end. The diameter of a piece having the same hardness in the centre is 80 mm when quenched in water, and 55 mm when quenched in oil.

remaining 13 measurements every 1.5 mm. For each two measurements, taken at opposite sides, the arithmetical mean is calculated.

The hardenability test results for the steel are represented in a graphical form (Fig. 40). From the hardenability test results obtained

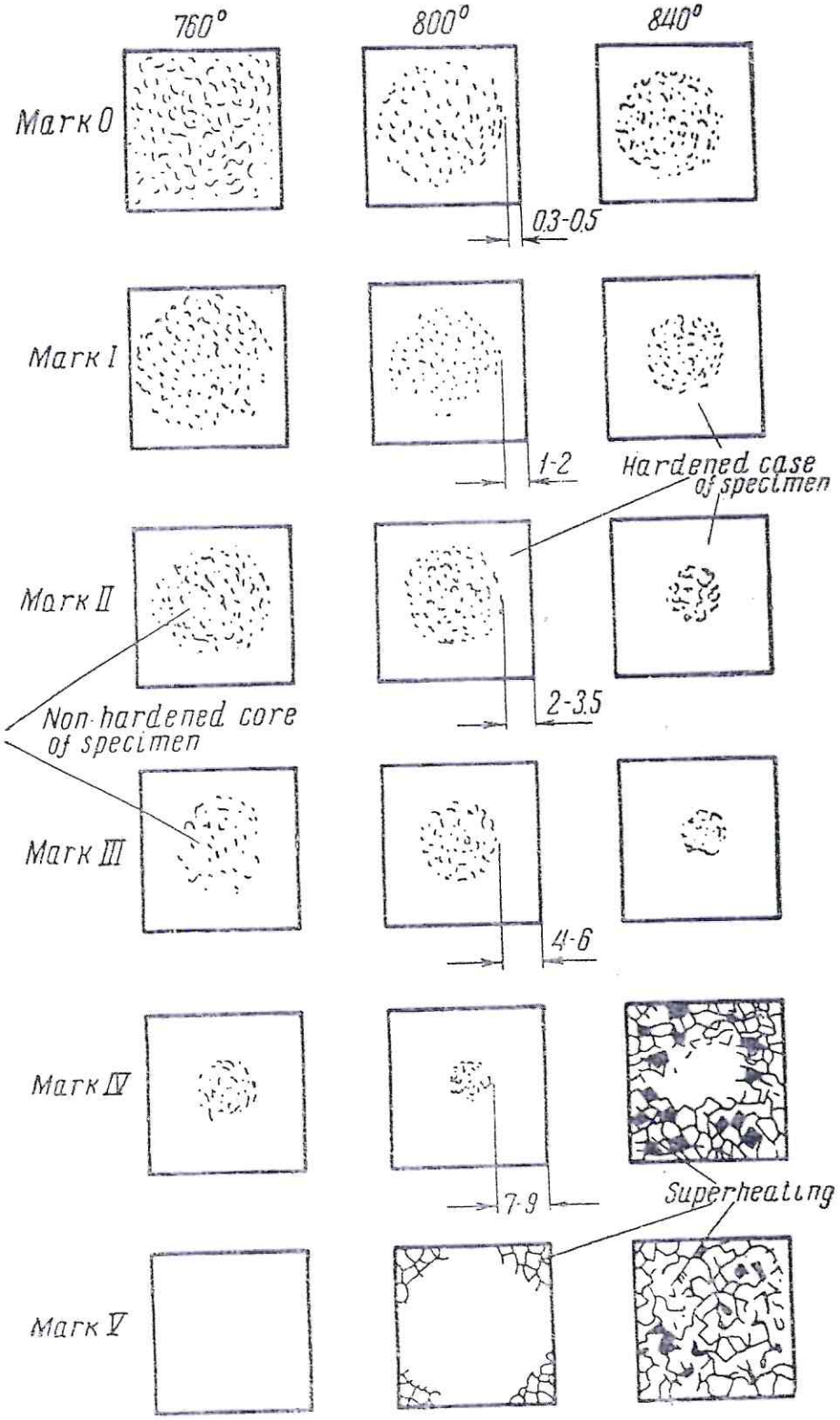

Fig. 42. The hardenability scale for carbon tool steels (according to State Standard).

by the end-quenching method, the diameter of the cylindrical part, having the same hardness in the centre when quenched either in water of temperature 10 to 20° or in oil with temperature of 15 to 70°, may be determined, the evaluation being made on the basis of the diagram shown in Fig. 41.

The evaluation of the hardenability of carbon tool steels is made in accordance with the so-called fracture scale (Fig. 42) specified by the U.S.S.R. State Standard. The hardenability of the steel is estimated according to the fractures of three specimens having a cross-section of 20 by 20 mm, or round specimens 21 to 23 mm in diameter and 100 mm long, notched at the centre (the depth of a notch being 2 to 5 mm, or 5 to 7 mm for key-hole type). The specimens are first annealed at a temperature range of about 730 to 750° and then, after being held for 2 hours at this temperature, they are cooled with a furnace down to 650°C. They are then quenched in water of 10 to 30° at temperatures of 760, 800, and 840°C.

19. Cooling Media

Various liquid coolants are used in quenching steels as the principal quenching (or cooling) media.

Let us consider the process of cooling by immersion in a cooling liquid an article which has been heated to a high temperature. When it is immersed in a liquid coolant the liquid layer immediately surrounding it will almost instantly be heated up to boiling point and vaporised. The article will be enveloped in a steam layer or vapour coating, which will prevent further access of the cooling liquid to the surface of the article. The surface layers of the article will be cooled very intensively before the formation of the vapour coating.

To evaporate a liquid substance it is necessary:

1) to heat the liquid from its original temperature t_0 to boiling point t_{boil}; the amount of heat required for this purpose is proportional to the specific heat of the liquid c_{liq} and temperature difference $t_{boil}-t_0$:

$$c_{liq}(t_{boil}-t_0);$$

2) to heat the liquid up to boiling point t_{boil} in order to change it into the gaseous state; the amount of heat required for this purpose is proportional to the latent heat of vaporisation of the liquid L_{vapour};

3) to heat the vapour from the boiling point of the liquid to a certain final temperature t_f; the amount of heat required for this purpose is proportional to the specific heat of vapour c_v and temperature difference t_f-t_{boil}:

$$c_v(t_f-t_{boil}).$$

The vapour coating formed is continuously washed off by the surrounding layers of liquid, this effect being the more pronounced the less the viscosity of the liquid, i. e., the greater its fluidity.

When an article is cooled below the boiling point of the liquid the vaporisation process comes to an end, and the heat of the article is transferred to the liquid through thermal conductivity only, the further cooling being the more drastic, the higher the heat conductivity.

So the quenching power or withdrawal of heat of a cooling liquid depends upon: a) its initial temperature; 2) its boiling point; 3) the specific heat of the liquid; 4) the latent heat of vaporisation; 5) the specific heat of vapour; 6) the thermal conductivity; 7) viscosity.

The quenching media most widely used in commercial practice are water and mineral oils. The cooling efficiency of the mineral oils is by many times less than that of water because their specific heat, latent heat of vaporisation and thermal conductivity are lower than those of water, while the viscosity is considerably higher, as is evident from Table 5 compiled by D. Y. Vishnyakov.

Table 5

Comparative Specifications for Physical Properties of Quenching Media
(values for water are assumed to be 100 per cent)

Quenching liquid	Boiling point	Latent heat of vaporisation (total)	Thermal conductivity	Specific heat	Viscosity	Hardness at the surface of quenched steel
Water	100	100	100	100	100	100
Machine oil of grade JI	330	45	17	46	440	49
Spindle oil of grade H . . .	330	45	17	48	320	49

In order to obtain the martensitic structure in hardening carbon and most low-alloy steels, conditions should be created which enable the quenching to be carried out at a very high cooling rate (faster than the critical rate) through the range of Ar-transformation, where the stability of the austenite is at its lowest value (650 to 550°); this prevents the transformation of austenite into a ferrito-carbide conglomerate. On the other hand, it is desirable to cool an article at a considerably slower rate within the range of martensitic transformation (300 to 200°). In this case the internal thermal stresses, which invariably develop, are bound to be correspondingly lower, which is especially important for the steel in this temperature range since its ductility decreases drastically. A liquid that would permit